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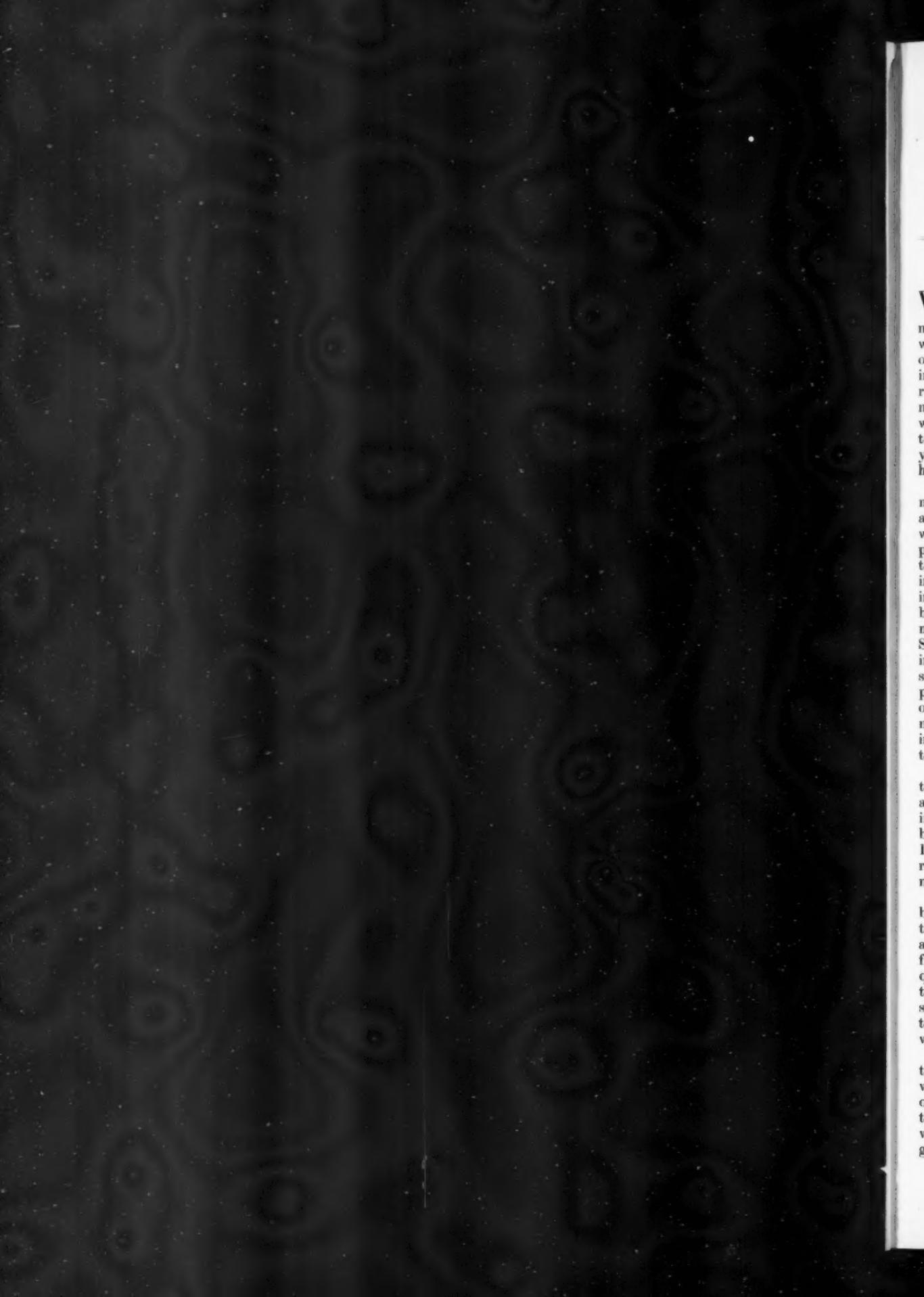
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Steel Progress

WHEN the steel industry's development plan was reviewed in 1948 in consultation with the Government planning authorities, it was estimated that demand would reach 18 million tons by the mid-fifties. The output for 1953 was of the order of 17½ million tons and in the last quarter of the year production was already running at an annual rate of 18½ million tons. In every month of the year, the output of ingots and castings was above the 1952 level, and the gain over the whole twelve months was the largest recorded in any post-war year except 1948, when production rose rapidly after having been seriously affected by the fuel crisis of 1947.

This increase of well over a million tons in 1953 was made possible by the increased capacity available as another stage in the industry's development programme was completed. As scrap for steelmaking is now less plentiful than it was in the immediate post-war years, the expanding steel output must necessarily be based on increased pig iron production. During the year, the industry has had the benefit of the six new or rebuilt blast furnaces blown in during 1952. Early in 1953, moreover, the first of the new furnaces at the Shotton plant of John Summers—the largest built so far in this country—came into operation, while less spectacular but substantial extensions of capacity took place elsewhere. The total production of all qualities of pig iron rose about half a million tons to appreciably more than 11 million tons, but as the demand for foundry iron was less active than the previous year, the extra tonnage available for steelmaking was rather larger.

The rise in steel production was more than sufficient to cover the increase in home and export requirements, and it was, therefore, possible to make a substantial cut in imports of ingots and semi-finished steel, which had been increased abnormally in 1952. The substantial 1953 imports—a little over a million tons—are a reflection of the extent to which consumers and merchants have increased their stocks.

The statistics, showing as they do an overall balance between supply and demand, might lead to the conclusion that the consumers' worries are now a thing of the past, and that everything in the garden is lovely. Apart from the fact that the right type of steel in the right quality is not always available in just the right quantity to meet a customer's requirements, and that a small shortage can create apparent chaos out of all proportion to its size, there have been two classes of product for which supply and demand have not been in balance.

There is still a "hump" in the demand for plates—that is, there is an excess of temporary demand over what the steel industry believes to be a reliable estimate of the long-term demand. The "hump" is ascribed to two main causes; the catching up of arrears in railway wagon building, and an intensified shipbuilding programme consisting in large part of tankers, which require

more plate per gross ton than dry cargo vessels. The "hump" is expected to level itself out, and the provision of extra plate rolling capacity could only lead to surplus, and increased overhead costs. Moreover, any extra capacity planned now would probably not be available before the "hump" disappeared. In these circumstances it is essential that every effort should continue to be made to utilise to the full the existing facilities. The 1953 output of plates at around 2½ million tons is 10% up on the 1952 figure and 67% greater than the 1937-8 average. As a consequence of slightly increased imports and slightly reduced exports, the supplies to the home market have been about 12½% above the 1952 level.

The installation of new capacity brought sheet supply into balance with demand in early 1952, with the result that sheet imports had virtually ceased by the end of the year. Early in 1953 home demand subsided somewhat and arrangements were made to increase exports to utilise available capacity to the full. Later in the year the demand again hardened, but as a result of new capacity a further increase in sheet production should be available next year.

It is evident, therefore, that when the supply of and demand for steel are in general in balance, the demands for particular steel products may still have their "humps." One way of overcoming the difficulty would be to have surplus manufacturing capacity for each class of product, but this would, of course, mean increased overheads—and prices—due to the excess capacity remaining unused in normal times. In the view of the steel industry, the best way of dealing with the humps would seem to be, not through the provision of excess mill capacity; nor through violent fluctuations in exports, with the consequent disturbance in relations with foreign customers; but through variations in the import programme. It is for this reason that since 1935 an import in normal times of half a million ingot tons has been envisaged.

Further additional blast furnace and steel furnace capacity will come into production in 1954. On the pig iron side, two large new furnaces will be blown in at the Appleby-Frodingham works—one within a few weeks and the other later in the year. At Richard Thomas and Baldwin's works at Scunthorpe and Ebbw Vale, the installation of additional equipment will, in each case, enable three furnaces to operate simultaneously, instead of two as at present; and smaller increases will occur elsewhere. On the steel side, the development of the duplexing process at Consett (acid Bessemer and basic open hearth) will mean a considerable increase in steelmaking capacity there; a new 60-ton electric furnace will be completed at Samuel Fox's; and there will be extensions at Ebbw Vale, Bilston and other works. Arrangements have been made for a supply of home and imported ore, and there is every assurance that the extra coking coal required by the industry will be available.

February Diary

1st

Institute of British Foundrymen (Sheffield and District Branch). "Modern Methods of Pattern Making with Special Reference to American Practice and Pressure Cast Aluminium Pattern Plates," by B. N. S. PERRY. Sheffield College of Commerce and Technology, Department of Engineering, Pond Street, Sheffield, 1. 7.30 p.m.

2nd

East Midlands Metallurgical Society. Members Night. Short papers to be given by Society Members. Nottingham and District Technical College, Shakespeare Street, Nottingham. 7.30 p.m.

Institute of Metals—Oxford Local Section. Junior Members' Evening. Discussion on New Topics. Oak Room, Cadena Cafe, Cornmarket Street, Oxford. 7 p.m.

3rd

Institute of Fuel (Yorkshire Section). "Fuel Oil and Oil Firing," by G. J. GOLLIN. The University of Sheffield. 6.30 p.m.

Institution of Engineering Inspection. "The Organization of Inspection on the Lines of Statistical Quality Control," by J. FORSTER-COOPER. Room 3, Birmingham Chamber of Commerce, New Street, Birmingham. 7.30 p.m.

Institution of Engineering Inspection. "Quality Control in the U.S.A.," by S. W. NIXON. Room A5, Coventry Technical College. 7.30 p.m.

Manchester Metallurgical Society. "Solidification in Castings," by R. W. RUDDLE. Lecture Room, The Central Library, Manchester. 6.30 p.m.

4th

Institute of Metals—Birmingham Local Section. "The Fracture of Steel," by J. E. RUSSELL. James Watt Memorial Institute, Great Charles Street, Birmingham. 6.30 p.m.

Institute of Metals—London Local Section. "Germanium and Silicon," by R. W. DOUGLAS. 4, Grosvenor Gardens, London, S.W.1. 6.30 p.m.

Institution of Engineering Inspection. "Copper and its Applications, with particular reference to the Activities of the Copper Development Association," with film, by G. W. PRESTON. Royal Society of Arts, John Adam Street, Adelphi, London, W.C.1. 6 p.m.

Leeds Metallurgical Society. "Nucleation in Metals and Alloys," by J. H. O. VARLEY. Chemistry Department, The University, Leeds, 2. 7.15 p.m.

8th

Institute of Metals—Scottish Local Section. "Some Practical Hints on the Production of Non-Ferrous Castings," by WILLIAM DUNLOP. Institution of Engineers and Shipbuilders in Scotland, 39, Elmbank Crescent, Glasgow, C.2. 6.30 p.m.

Institute of Welding (Sheffield and District Branch). "Developments in Welding Research," by Dr. A. A. WELLS. Sheffield College of Commerce and Technology, Pond Street, Sheffield, 1. 7.15 p.m.

Institution of Production Engineers (Sheffield Section). "Recent Research Developments in Engineering," by Dr. T. E. ALLIBONE. The Grand Hotel, Sheffield. 6.30 p.m.

North East Metallurgical Society. "Protection of Steelwork from Atmospheric Corrosion," by Dr. F. R. HIMSWORTH. Cleveland Scientific and Technical Institution, Middlesbrough. 7.15 p.m.

9th

Institute of Metals—South Wales Local Section. "The Zinc Die-Casting Alloys," by L. A. J. LODDER. University College, Metallurgy Department, Singleton Park, Swansea. 6.45 p.m.

Sheffield Metallurgical Association. "High Temperature Insulation," by G. MARSH. B.I.S.R.A., Hoyle Street, Sheffield. 7.30 p.m.

11th

Incorporated Plant Engineers. "Metallurgy," by J. ROLSTON. Roadway House, Oxford Street, Newcastle-on-Tyne. 7 p.m.

Liverpool Metallurgical Society. "Diffusion in Metals," by A. D. LE CLAIRE. Liverpool Engineering Society, The Temple, Dale Street, Liverpool. 7 p.m.

12th

Institute of Fuel (Yorkshire Section). "Aspects of a National Fuel Policy," by G. NABARRO, M.P. The University of Leeds. 6.30 p.m.

15th

Sheffield Society of Engineers and Metallurgists. "The Future of Steel Melting," by PROF. W. M. THRING. University Building, St. George's Square, Sheffield. 7.30 p.m.

16th

Institute of British Foundrymen—East Anglian Section. "Outlook and Attitude of Management and Workers of British and American Foundries," by R. F. HORTON. Central Hall, Public Library, Ipswich. 7.30 p.m.

Institute of British Foundrymen—Slough Section. "Runners and Risers," by J. WOOD. Lecture Theatre, High Duty Alloys, Ltd., Slough. 7.30 p.m.

Sheffield Metallurgical Association. "Analytical Chemistry of Boron," by H. G. SHORT. B.I.S.R.A., Hoyle Street, Sheffield. 7 p.m.

17th

Manchester Metallurgical Society. "New Optical Techniques," by Dr. G. K. T. CONN. Lecture Room, The Central Library, Manchester. 6.30 p.m.

18th

Society of Chemical Industry—Corrosion Group. "The Use of Non-Ferrous Metals in Domestic Water Supply," by H. S. CAMPBELL. Joint Meeting with The Road and Building Materials Group. Institution of Structural Engineers, 11, Upper Belgrave Street, London, S.W.1. 6 p.m.

19th

West of Scotland Iron and Steel Institute. "Notch Ductility of Quenched and Tempered Steels," by S. DOWNS; "Vertical Rolling Mills," by J. A. KILBY and G. A. V. RUSSELL. 39, Elmbank Crescent, Glasgow. 6.45 p.m.

23rd

Institute of Fuel. "Carbonization of Blends of Coals to Produce Metallurgical Coke," by H. BARDGETT. Institution of Mechanical Engineers, Storey's Gate, St. James's Park, London, S.W.1. 5.30 p.m.

Institution of Engineers and Shipbuilders in Scotland. "Steel Wire Ropes: their Construction and Application," by F. J. HEWITT. 39, Elmbank Crescent, Glasgow. 6.45 p.m.

Sheffield Metallurgical Association. "25 Years of Co-operative Research on Corrosion," by DR. J. C. HUDSON. Joint Meeting with the Royal Institute of Chemistry. B.I.S.R.A., Hoyle Street, Sheffield. 7 p.m.

24th

Institute of British Foundrymen—London Branch. "Propeller Manufacture," by J. M. LANGHAM. Waldorf Hotel, London, W.C.2. 7.30 p.m.

25th

Institution of Engineering Inspection. "Testing of Metals with Mechanical Vibration," by DR. K. ENTWISTLE. The Engineers Club, Albert Square, Manchester. 7.30 p.m.

26th

Incorporated Plant Engineers. "Tacoma Bridge Failure," by B. G. OUGH; and "Fatigue Fractures," by J. LANG. Imperial Hotel, Birmingham. 7.30 p.m.

Institute of Metals—Birmingham Local Section. All-Day Symposium. Details to be announced. Birmingham College of Technology.

Evolution of Hydrogen from Weld Metal

By K. Winterton, Ph.D., B.Sc., A.I.M., A.M.I.W., and
C. L. M. Cottrell, Ph.D., M.Sc.

British Welding Research Association

Hydrogen evolving at room temperature from freshly made steel weld metal can be measured by a simple displacement method described by the authors. Values of evolved hydrogen are quoted for 19 electrodes of various kinds. Experiments have also been made to determine the effect of welding current, storage and baking conditions, and the plate surface condition on the amount of hydrogen evolved. The importance of hydrogen content in relation to cracking problems is briefly discussed.

IN view of the important influence of the hydrogen concentration in the weld metal on hardened-zone cracking when welding alloy steels by the metal arc process, it was thought desirable to compare various electrodes from this point of view. Recent work has fully established the theory put forward by Hopkin,¹ that cracking in the hardened zone is due to diffusion of hydrogen from the weld metal, combined with the severe martensite transformation stresses.

In a recent paper by Rollason and Roberts,² a description was given of an unusual form of cracking in mild steel weld metal, which occurred at intermediate temperatures: this is not to be confused with the more usual type of cracking in weld metal—hot cracking—which occurs at much higher temperatures. The new form of cracking occurs only under severe cooling conditions, and was attributed to the effect of hydrogen concentration, acting in combination with the transformation stresses in the weld metal. The explanation is similar to that now given for hardened zone cracking, though, since the transformation occurs at higher temperatures, more severe conditions are necessary to obtain cracking. More recently, some thought has been given to the possibility that hydrogen concentration may play a part in hot cracking in weld metal.

The extension of the influence of hydrogen to cracking in weld metal entailed in the new ideas adds further point to the collection of comparative data for various electrodes and welding conditions.

Method

A single bead of weld metal, 2 in. in length, was laid along the centre of a small mild steel block, 2 in. \times 1 in. \times $\frac{1}{4}$ in. The weld was laid from a marked 3-in. length of electrode, and, whenever possible, an 8-gauge electrode was used at a standard welding current of 185 amp. Details are given later of any deviations from this technique. As soon as the weld was made, the block was quenched in water, then cleaned from slag and dried. It was then placed under an inverted funnel in a beaker of boiled* paraffin. These operations were conducted as rapidly as possible, the time elapsing between the end of welding and the immersion of the specimen in paraffin being of the order of 10 seconds. The throat of the funnel was connected with the open end of an inverted burette (2 ml. capacity) in which the evolved gas was collected. This arrangement is shown in Fig. 1, and a

closer view in Fig. 2 shows the hydrogen evolution in progress. This method is similar to that used by Andrew and co-workers.³

To measure the volume of gas evolved, the level of paraffin in the burette was first brought to a position just within the calibrated part of the tube. This meant that the evolved gas mixed with air at the top of the tube. The volume of evolved gas could then be estimated from the difference in levels at the beginning and end of gas evolution. Slight corrections had to be made for the effects of temperature and pressure on the gas volume. Readings were usually taken at 20°C., but when this was not possible, a small correction was made. Changes in temperature influenced both the change in volume of the gas and also the vapour pressure of the paraffin; the extent of the correction was, therefore, determined experimentally by noting the changes in level in a column of paraffin as the temperature was varied. A second correction must be applied for the change in pressure to which the gas is subjected as the level of paraffin falls.

In preliminary experiments, ordinary fuel paraffin was used, but it was found that hydrogen was soluble in this liquid. Medicinal paraffin, a higher fraction with greater viscosity, was substituted and found satisfactory, having a low solubility for the gas.

An important feature of the method is that a very short single run weld sample is used, involving a deposit weight of about 7 g. This allows a minimum opportunity for hydrogen to escape while the sample is being made. In particular, it was thought desirable to avoid building up a multi-run deposit, since the heat-treatment cycle involved in making each run must speed up the escape of hydrogen. The results showed this procedure to be justified, in that much higher values for hydrogen concentration were recorded than have been previously obtained. The use of a small sample had the further advantage that evolution of gas was substantially complete in a matter of a few days. It should be noted that, in practice, short single-run welds (particularly fillets) are commonly used, and that this is usually found to be the most critical condition for cracking to occur, either in the weld or in the adjacent parent plate. After the hydrogen estimation, the weld samples were sectioned for examination, and the melted area was measured by means of a planimeter. Any porosity present was also measured in the same way. In addition, the specimens were weighed before and after completion of the test to give the weight of the deposited or added metal.

* The paraffin was boiled before use to get rid of all traces of dissolved gas, which otherwise might affect the results obtained.



Fig. 1.—Apparatus used for measuring hydrogen evolution.

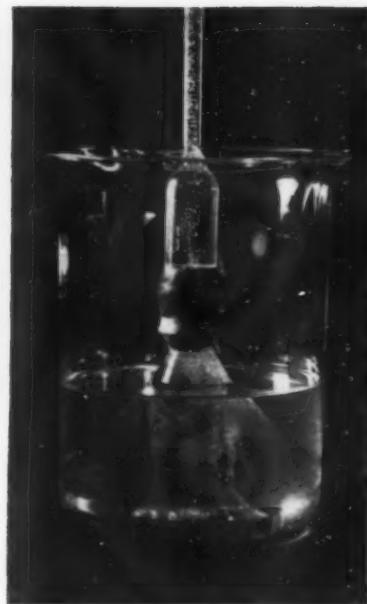


Fig. 2.—Near view showing evolution of hydrogen.

Results

Hydrogen Content for Various Electrodes

The gas which is evolved at room temperature from mild steel weld metal is practically pure hydrogen. This fact has been established elsewhere by careful gas analysis. Tests have been made to confirm this by adsorbing the evolved gas on palladised asbestos; the results showed that the gas was completely adsorbed. In the present paper, the evolved gas has, therefore, been referred to as hydrogen. A test made with a bare wire electrode gave a very low value for evolved gas (Test S in Table I); with no coating, the only available moisture for producing hydrogen is atmospheric water vapour, together with any water associated with traces of oxide on the electrode and plate metal.

A preliminary survey has been made of various electrodes from the point of view of the amount of hydrogen evolved at room temperature. The results are set out in Table I. Two sets of figures are given for the hydrogen contents in the tables; the first is worked out on the basis of the density of hydrogen in the deposited or added metal, whilst the second refers to the density in the weld, i.e. in the melted zone. The two sets of figures give practically the same relative order, but the second figure is probably the more significant.* A separate code letter has been reserved for each electrode, and later, when further tests are made using the same electrode, they are coded, for example, B2, B3, etc. In Table I the electrodes are presented in groups according to the coating type, i.e. according to the first number in the B.S.1719 classification.

* Recent tests in these laboratories, in which the hydrogen content evolved at room temperature from a steel sample has been determined in a gas analysis apparatus, indicate that the figures recorded here may be somewhat low (perhaps by as much as 25%), presumably due to solution of hydrogen in the paraffin. In this connection, it has been remarked that higher figures can usually be obtained by displacement over mercury. However, the main features remain unaltered, and the values obtained by the method reported here are comparative for the various electrodes, since every attempt was made to ensure that conditions were kept comparable.

The first important point is that in many instances very high values for evolved hydrogen have been obtained, compared with the recorded values of previous workers. This is thought to be due to the use of a small single-run weld to ensure minimum loss of hydrogen in making the specimen. Presumably, these higher figures have a greater significance from the point of view of the dangerous diffusible hydrogen associated with a given electrode.

The highest individual value for evolved hydrogen was given by an electrode of the cellulosic type (Class E 111 P). It should be stated that this electrode could not be given the standard prior baking treatment of 5 hours at 110°C., since this treatment impairs the welding quality. However, the result is as expected, since cellulose would be expected to yield a good deal of water vapour, from which hydrogen is formed. High values have also been obtained for two of the electrodes of the rutile-coated type, and, in general, all electrodes tested in

the first four groups seem to give dangerously high values.

Six electrodes have been tested of the lime-ferritic or hydrogen-controlled type (Tests K—P), and the results show fairly low and consistent values. Four out of six electrodes tested in this class showed porosity and, no doubt, some of the hydrogen becomes trapped in these pores in the molecular condition. Low values of hydrogen were found in the oxide-coated electrodes, particularly with Electrode I, and this is in line with their freedom from cracking in the weld and adjoining plate material, even with difficult steels (high-tensile and free-cutting). No gas was evolved from either of two austenitic electrodes tested. The success of austenitic electrodes in the welding of alloy steels has been attributed to the fact that the hydrogen is held in solution in the austenite, instead of diffusing into the hardened zone of the plate material. Strictly speaking, this is only a pseudo-solubility, being due to an exceptionally low permeability for hydrogen in austenite at room temperature. (See Rollason⁴).

As might be expected, a low value was obtained with a bare wire weld, since the only available source of hydrogen is the atmospheric water vapour, possibly augmented by moisture associated with traces of oxide present on the wire or plate surfaces.

Effect of Welding Current

A series of tests was made with a rutile-coated electrode (E 217), in which the welding current was varied over a wide range. The results are given in Fig. 3, and show clearly that the amount of hydrogen evolved increases considerably with increasing welding current. It was also observed that, as the current was raised, the deposit efficiency was brought down from the initially high value of 92% to about 80%, and in one special case at 230 amps. the yield was only 72%.

Hydrogen Evolution in Cold Weather

In all tests, an attempt was made during the day to keep the temperature to a value of about 20° C., but this was not always possible for the whole of the day. At night, in very cold weather the temperature would fall to 10° C. or below. Where this occurred, the evolution of hydrogen seemed to be considerably delayed, so that in some cases 23 days elapsed before completion. Under these conditions, there may be an opportunity for solution of some of the hydrogen in the paraffin, giving spuriously low readings. Experiment has shown that hydrogen left in contact with paraffin for long periods under these conditions does dissolve slightly. It was found that 0.04 ml. was absorbed in 14 days. By extrapolation, this would account for an error of 7-8% in readings taken after 23 days, though the effect would be negligible for periods of up to three days. Greater errors are to be expected in practice, due to the large surface area of the minute bubbles of hydrogen evolved.

The substantial effect of such a small change in temperature on the evolution rate is rather surprising, though a partial explanation is available from the following considerations. The diffusivity relation for gases in metals is given by :—

$$D = \frac{K}{d} p^m e^{-E_0/2kT}$$

where d = distance

p = pressure difference

T = absolute temperature

k = Boltzmann constant

K , E_0 and m = constants.

Substituting the appropriate values for the diffusion of hydrogen in α-iron, we obtain :—

$$D \propto e^{-\frac{15 \times 10^8 \times 4.2 \times 10^7}{2 \times 1.4 \times 10^{-23} \times 6 \times 10^23 \times T}} \text{ or } D \propto e^{-15000/4T}$$

TABLE I.—HYDROGEN EVOLVED AT ROOM TEMPERATURE FOR VARIOUS ELECTRODES

(Preliminary Baking Treatment 5 Hours at 110°C.)

Code	B.S. 1719 Class	Test Duration (days)	Hydrogen Evolved (ml.)	Deposit Weight (g.)	Hydrogen Density of Deposit ml./100g.	Hydrogen Density of Melt ml./100g.	Porosity (%)
A ...	E.111P‡	23*	2.36§	4.80	49	22	—
R ...	E.217	1	1.28	6.6	19	19	—
O ...	E.217	4†	2.59§	6.1	43	21	—
D ...	E.217	1	1.80	5.2	35	14	—
E ...	E.317	1	1.56	6.2	25	15	—
F ...	E.317	1	1.30	5.9	19	11	—
G ...	E.416	1	1.43	6.6	23	13	—
H ...	R.416	1	1.61	5.5	29	16	—
I ...	E.527	1	0.01	6.2	0.2	0.08	—
J ...	E.537	5	0.19	5.39	5.6	1.5	—
K ...	E.615	1	0.24	7.5	3.2	2.1	—
L ...	E.645	1	0.31	6.6	4.7	2.3	7.9
M ...	E.614	1	0.19	7.5	2.5	1.4	3.0
N ...	E.614	1	0.25	8.8	2.9	2.0	—
O ...	E.614	1	0.21	7.9	2.7	1.7	1.4
P ...	E.615	1	0.20	6.7	3.0	1.5	1.2
Q ...	18/8	1	0.00	7.3	0.0	0.0	—
R ...	18/8	1	0.00	7.3	0.0	0.0	—
S ...	Bare Wire (ferritic)	1	0.03	5.9	0.5	0.4	2.6

* This test was made in exceptionally cold weather; evolution was 84% complete in one day.

† This test was made in cold weather, evolution was 91% complete in one day.

‡ This electrode was tested after storing in a warm cupboard. Baking, even at 110°, impairs the welding quality with this class of electrode.

§ Though the burette used was of 2 ml. capacity, larger volumes of gas can be measured by readjustment of the liquid level at some point during evolution.

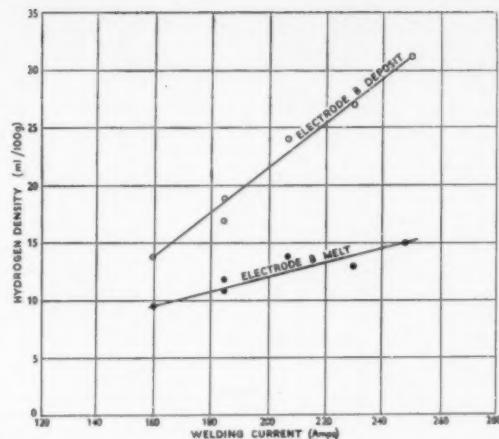


Fig. 3.—Effect of welding current on the evolution of hydrogen.

On substituting the appropriate values of T , we find that, by reducing the temperature from 27° C. to 17° C., the diffusivity is reduced by approximately 56%.

Effect of Storage Conditions and Baking Treatment on Hydrogen Evolution

Comparative tests have emphasised the importance of proper storage conditions, the relevant data being given in Table II. Electrodes of three different kinds were stored for 34 days in a very moist atmosphere before testing. An electrode of the rutile-basic type (E 317) seemed comparatively insensitive to this treatment, the hydrogen density (deposit) being 26 ml./100 g. (Test E 2, Table II) compared with 25 ml./100 g. for the electrode in the baked condition (Test E, Table II). On the other hand, with two electrodes of the lime-ferritic type, the hydrogen density increased to dangerous proportions. (Compare Tests L 2 and N 3 with tests L and N in Table II). This is not surprising, if we assume that lime-ferritic electrodes owe their low hydrogen content to a high-temperature baking treatment, so that with this class subsequent storage conditions are particularly important to maintain their advantage.

It appears that electrode coatings vary considerably in their capacity to absorb water when stored in damp conditions, and it will be seen that Electrode N is superior to Electrode L in this respect.

The tests mentioned above involved storage of the electrodes for 34 days in a very moist atmosphere. This had a more deleterious effect, with one electrode, than complete immersion in water for a week. (Compare Tests N 2 and N 3, Table II).

For the comparison of various electrodes from the point of view of hydrogen evolution, a standard baking treatment of 5 hours at 110° C. was given before testing. A true baking treatment has a definite advantage over simply storing the electrodes in a warm dry place. For example, a rutile-coated electrode taken from a heated store cupboard (Test B 10, Table II) yielded a hydrogen density of 19 ml./100 g. (melt) compared with 12 ml./100 g. after baking at 110° C. A lime-ferritic electrode from the same store (Test L 3, Table II) yielded a hydrogen density of 5.0 ml./100 g. (melt) compared with 2.3 ml./100 g. Similarly with Electrode D, the value for hydrogen density is 18 ml./100 g., after warm storage, the figure being reduced to 14 ml./100 g. after

TABLE II.—EFFECT OF STORAGE CONDITIONS AND BAKING TREATMENT ON HYDROGEN EVOLUTION

Code	B.S. 1719 Class	Storage Condition	Test Duration (days)	Hydrogen Evolved (ml.)	Deposit Weight (g.)	Hydrogen Density of Deposit (ml./100 g.)	Hydrogen Density of Melt (ml./100 g.)	Porosity (%)
E2	E.317	Exposure to damp ^o	2	1.70	6.48	26	15	—
L2	E.645	" " " "	2	1.90	5.35	36	20	—
N3	E.614	" " " "	2	1.73	8.44	21	12	0.9
N2	E.614	" " " "	1	0.36	8.5	4.3	2.6	1.2
B10	E.217	Warm store cupboard	1	2.04	6.6	30	19	—
D2	E.217	" " " "	1	1.53	5.5	28	18	—
L3	E.645	" " " "	1	0.61	6.6	9.7	5.0	3.0
Q2	18/8	" " " "	1	0.00	7.1	0.0	0.0	—
B	E.217	5 hrs. at 110° C.	1	1.28	6.6	19	12	—
D	E.217	" " " "	1	1.80	5.2	35	14	—
E	E.317	" " " "	1	1.56	6.3	25	15	—
I ₂	E.645	" " " "	1	0.31	6.6	4.7	2.3	7.9
N	E.614	" " " "	1	0.25	8.8	2.9	2.0	—
B3	E.217	10 hrs. at 110° C.	7	1.04	6.21	17	11	—
B11	E.217	25 " " 110° C.	6	1.15	6.78	17	9.7	—
B12	E.217	2 " " 550° C.	1	0.47	6.5	7.2	4.9	—

^o Treatment of 34 days in a very moist atmosphere.[†] Immersed 7 days in water, afterwards storing for 1 day in a moist atmosphere.[‡] Stored in a dry cupboard at room temperature.

baking for 5 hours at 110° C. (It is surprising that the figures for deposit density are reversed in this particular case, and the point requires further investigation).

Treatment at 110° C. for longer than 5 hours did not have much further effect, as shown in Tests B 3 and B 11, Table II, the value for hydrogen density being reduced from 12 ml./100 g. after 5 hours to 11 ml./100 g. after 10 hours and 9.7 ml./100 g. after 25 hours.

Treatment at still higher temperatures—2 hours at 550° C.—brings a further marked reduction in hydrogen density, though it should be noted that in most cases this is not a practicable treatment, since the welding characteristics are affected.

Effect on Hydrogen Evolution of (a) Oil and (b) Scale on the Plate Surface

Two electrodes were chosen for these tests, one from the ordinary rutile-coated class and one from the lime-ferritic class.

To examine the effect of oil, the plate sample was coated with light machine oil before welding. The results are shown in Table III. With both electrodes, considerably less hydrogen was evolved than was normally the case, though with the lime-ferritic electrode, this was accompanied by an increase in porosity. It seems likely that this effect would be explained by the sudden evolution of hydrocarbons by decomposition of the oil, which would tend to shield the arc from water vapour.

Using the same electrodes, preliminary experiments have also been made to test the effect of oxide scale on the plate surface. For this purpose, the plate samples were heat treated for given periods at 900° C. On

cooling down, a period of about one hour was allowed to elapse before welding. With a light scale present on the plate surface, there was a considerable reduction in the amount of hydrogen evolved with the lime-ferritic electrode. There was a corresponding, though much smaller, reduction in the amount of hydrogen evolved from the rutile-coated electrode when using a light scale, but a considerable increase in hydrogen evolution for the heavily oxidised specimen. It seems likely that there are two competing effects with oxide scale. Oxide incorporated in the weld metal will tend to change its characteristics in the direction of the oxidising (dead-soft) type, causing a reduction in the hydrogen (and carbon) content in the weld metal. On the other hand, the presence of oxide scale permits the absorption of moisture which can increase the hydrogen content of the weld metal. Presumably the second effect becomes more important as the thickness of oxide coating is increased. Evidently further work is necessary to clear up these points.

Discussion

Hydrogen in Relation to Hot Cracking

Some work has been done to examine the effect of hydrogen on hot cracking. It was apparent that a rough correlation existed between the hot cracking tendency (measured in the B.W.R.A. double-fillet test) and hydrogen content. In addition, it was shown that, when the hydrogen was reduced by baking treatment, the cracking was diminished, and when the hydrogen was increased by humidifying treatments, the tendency to crack was likewise increased. However, the work has

TABLE III.—EFFECT OF CONDITION OF PLATE SURFACE ON HYDROGEN EVOLUTION

Code	B.S. 1719 Class	Plate Condition	Test Duration (days)	Hydrogen Evolved (ml.)	Deposit Weight (g.)	Hydrogen Density of Deposit (ml./100 g.)	Hydrogen Density of Melt (ml./100 g.)	Porosity (%)
M	E.614	Normal	1	0.19	7.5	2.5	1.4	3.0
M2	E.614	Oily	1	0.02	7.23	0.4	0.2	7.8
M3	E.614	Thin scale ^o	5	0.07	6.5	1.1	0.7	6.1
B	E.217	Normal	1	1.28	6.6	19	12	—
B13	E.217	Oily	5	0.77	6.65	12	7.6	—
B14	E.217	Thin scale [†]	5	1.15	6.39	18	12	—
B15	E.217	Thick scale [‡]	6	1.75	5.24	30	20	—

^o 15 minutes at 900° C.[†] 10 minutes at 900° C. yielding a total weight of 0.23 g. oxide.[‡] 30 minutes at 900° C. yielding a total weight of 0.72 g. oxide.

been of a preliminary character, and much remains to be done to establish more precisely the rôle of hydrogen with regard to this defect.

Hydrogen in Relation to Hard-Zone Cracking

From the results of hydrogen evolution tests on electrodes selected from the various coating groups (Nos. 1-6 in the B.S. 1719 code), it is possible to assess their relative merits with respect to liability to hard-zone cracking. In this connection, the most important factor appears to be the evolution of hydrogen during the transformation of austenite in the weld metal. The proportion of hydrogen forced out of solution at this change is considerable, since the solubility falls from about 5 ml./100 g. to half that value over a narrow range of temperature (see Fig. 4). The hydrogen which is evolved from the weld metal at this change diffuses under pressure into the heat-affected zone of the plate material, which will still be in the austenitic state if this material has a higher alloy content.

Hydrogen thrown out of solution during the solidification of the weld metal might also diffuse into the heat-affected zone of the plate material. However, since the results of tests with austenitic electrodes (Tests Q and R, Table I) have shown that no measurable amounts of hydrogen are liberated after welding, it must be assumed that the amount diffusing into the heat-affected zone during the solidification of the weld metal is negligible.

Because of the foregoing conclusion about the evolution of hydrogen during the solidification of the weld, it would appear that the hydrogen coming out of solution during the transformation of austenite is of primary importance for relating hydrogen evolution to hard-zone crack initiation. From this point of view, an electrode giving a greater evolution of hydrogen than 5 ml./100 g. of melted metal would be liable to cause trouble due to hard-zone cracking, this value being equivalent to the solubility limit at the start of transformation. Of course, the tendency to hard-zone cracking does not necessarily increase as the hydrogen content increases beyond the level of 5 ml./100 g.

Because of the sharp drop in solubility at the austenite transformation, it is to be expected that electrodes giving a hydrogen evolution of less than 5 ml./100 g. of melted metal will be less liable to cause hard-zone cracking as the hydrogen content is reduced. This suggestion is in agreement with the results obtained from the low-hydrogen electrodes (Group 6), which are much less liable to cause hard-zone cracking than electrodes from Groups 1, 2, 3 and 4. All the Group 6 electrodes have given a hydrogen evolution which is less than the hydrogen solubility at the end of the transformation of austenite, i.e., less than 2.5 ml./100 g. of melted metal. It should be noted that the oxidising electrodes (Group 5) also yield less hydrogen than this limit of 2.5 ml./100 g., and this type of electrode has proved in the past to be less liable to cause hard-zone cracking than electrodes of Groups 1, 2, 3 and 4.

In view of these considerations, the importance of keeping the low-hydrogen type of electrodes dry cannot be over-emphasised, since a little additional moisture would bring these electrodes into the danger region (2.5-5 ml./100 g. hydrogen evolution).

Acknowledgments

The authors would like to express their appreciation of the valuable discussions on the work given by members of the FM.3 and FM.8 Committees of the British

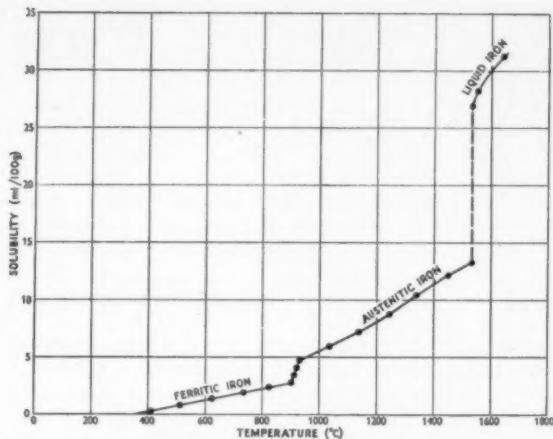


Fig. 4.—Solubility of hydrogen in iron at 1 atmosphere (Sieverts & Martin).

Welding Research Association. Special mention should be made of the helpful advice and interest of Dr. H. G. Taylor, Director of Research.

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Chair of Metallurgy at Nottingham

THE University Court was informed at its meeting on 16th December, that the Council of the University had that day gratefully accepted from Mr. C. T. Cripps, M.B.E., Chairman and Managing Director of Pianoforte Supplies Ltd., Simplex Works, Roade, Northampton, a gift of £100,000 for the endowment of two Professorial Chairs, a Cripps Chair of Metallurgy, and a Cripps Chair of Production Engineering.

Pianoforte Supplies Ltd. was founded by Mr. C. T. Cripps to produce components for the music industry. Whilst still retaining the same trading title, the firm, under his guidance and direction, has progressed into full scale production engineering, supplying numerous different industries with metal components, in all kinds of finishes. The Company realises the pressing need for trained personnel in the type of industry in which it is engaged, and it feels there is a great need for industrial development and research to keep this country in the front rank of engineering production.

The University of Nottingham has appealed to industry in the Midlands for support in its schemes to provide the necessary establishment to take care of engineering studies and research. As a result of this approach, the Company found that the aims of the University of Nottingham and the outstanding progress it had already made, were ideal for taking up two subjects in which the Company is practically engaged.

It is intended that an appointment to the Cripps Chair of Metallurgy should be made for the session beginning in October, 1954, when a degree course in metallurgy will be instituted as had already been planned. The Senate had previously given approval in principle to a proposed degree course in production engineering, but the date from which the Chair will be established will now be a matter for consideration by the Council.

Sintered Refractory Alloys

By A. Carter, M.A., A.I.M.
Hard Metal Tools, Ltd.

Increased gas turbine operating efficiencies can only be achieved by the use of higher operating temperatures, and the search for materials capable of service at such temperatures led to a study of the metal-refractory alloys discussed here. Although these materials are not yet in a position to replace the conventional forged blading alloys, considerable progress has been made. In this survey, particular reference is made to the alloys based on titanium carbide.

DURING the last ten to fifteen years a new series of materials suitable for high-temperature service has been introduced. These materials are alloys formed between a metal and a refractory constituent, and components are produced by the well-known techniques of powder metallurgy. Many combinations of metal and refractory are now being investigated, the metals commonly used being cobalt, nickel, iron, chromium, silicon, beryllium, titanium, molybdenum, tungsten or niobium. The refractory constituents are usually metallic carbides, nitrides, borides, silicides, or oxides. These alloys have been variously referred to as "ceramals," "cermets," "metamics," or other combinations of the words "metal" and "ceramic." Strictly speaking, these terms are inappropriate when applied to alloys containing carbides, nitrides, borides or silicides, as these materials are not true "ceramics," but have definite metallic characteristics, such as high thermal and electrical conductivities, and ability to alloy easily with metals. Alloys containing metal oxides, such as alumina, beryllia or magnesia are true "ceramals." Similar alloys have been in large scale production for many years, e.g. the cemented tungsten carbide tool materials, but these are not suitable for regular use at high temperatures owing to their poor oxidation resistance, and their high density, which makes them unsuitable for rotating components.

Physical and Mechanical Properties

The physical and mechanical properties of these metal-refractory alloys are often considered to be half-way between those of the metal and those of the refractory component. Both components play their part in determining these properties, since in most of these alloys both metal and refractory phases are present as nearly continuous skeletons. In general, the properties of these materials are based on the high resistance to creep and general high-temperature stability of the refractory phase, combined with the good thermal shock resistance, good thermal and electrical conductivities, and ductile character of the metal bonding phase. Many properties, such as high-temperature tensile creep resistance, depend to a large extent on the strength and structure of the interface between the metal and refractory constituents, and several types of metal-refractory bond are observed. Carbides, for instance, generally form a very strong bond as a result of solution of some carbide in the metallic bond during sintering to form a liquid "cementing" phase. The carbide is then usually reprecipitated from the bond during cooling. Alumina forms a "chemical" linkage with iron and chromium by the formation of a spinel-type double oxide interfacial bond. Other pairs of constituents form mechanical

bonds, relying on an interlocking formation for high strength.

Certain physical properties vary from type to type, depending on the particular alloy: for example, alloys based on metallic carbides have high thermal and electrical conductivities, and high thermal shock resistance, but those containing alumina or beryllia have lower properties. Oxide-base alloys have high oxidation resistance, but those based on carbides have lower resistance. Resistance to other types of corrosion and scale attack, such as vanadium pentoxide erosion, will vary considerably according to composition. The room temperature ductility of these materials varies from type to type, but is generally low, and when measured as an elongation in a tensile test or a bending in a transverse test is practically nil. Titanium carbide alloys bonded with large amounts of metal have the highest ductility, while the oxide-base ceramals are much more brittle. The high-temperature ductility is naturally much higher, and total elongations of 5-8% are observed on stress-rupture test pieces of nickel-bonded titanium carbide alloys at 980° C. (1,800° F.). These high-bond titanium carbide alloys have negligible ductility, as measured by elongation in a room temperature tensile test, but can be deformed locally to a surprising extent before cracking.

Uses of Metal-Refactory Alloys

The use of these metal-refractory compositions for turbine blading is a particularly attractive application, as they are very hard, and, consequently, have good resistance to erosion by dust particles; have lower densities than conventional forged blading alloys, and hence operate at lower centrifugal stresses; and have good resistance to thermal shock, creep and oxidation. Several successful short-run engine tests have already been performed with metal-refractory turbine rotor and stator blades. These alloys are used for other components, such as high-temperature furnace parts, tools for hot spinning, hot extrusion dies and mandrels for steel and light alloy extrusion, thermocouple protection sheaths, resistance heating elements, dies for hot compacting metal powders, and for many applications where a combination of hardness, wear resistance and corrosion resistance is required.

Methods of Production

In general, components are produced in these alloys by the usual well established methods of powder metallurgy, in particular, those by which cemented tungsten carbide shapes are produced. These methods can be divided into two distinct main processes, namely (1) cold pressing, followed by sintering, and (2) combined

hot pressing and sintering. If cold pressing methods are used, components are usually made from simple shapes which have been cold pressed and then given a low-temperature "pre-sinter," after which the pressings can be machined to shape using conventional machining processes, making due allowance for the contraction (sometimes as high as 30% linear), which will occur during final sintering. Metal-refractory mixes cannot be cold pressed to form complicated shapes, as the powders used are extremely fine and have poor flowing properties under pressure; only certain simple parts can be pressed to finished shape. Powders can be "plasticised" by suitable additions to the powder, and this enables slightly more complicated shapes to be pressed. The technique of hydrostatic pressing has recently been introduced, and by this process, irregularly shaped pressings are produced by placing the powder in suitably sealed rubber bags, and immersing these in a hydraulic cylinder. Using this method, very high pressures can be exerted equally over the entire surface of the pressing. Other cold forming methods include the extrusion of plasticised powders, the specialised technique of slip casting metal-oxide mixes, and the more recent use of a "cold casting" technique.¹ By means of the last named method certain powders can be "liquidised" with a liquid organic compound and poured into an intricately shaped wax mould, in which the organic compound polymerises into a solid at room temperature, or slightly above, and "freezes" the mixture into a solid shape. The wax mould is then removed by melting and the organic compound is volatilised by very slow heating.

The shapes produced by this whole series of cold-forming processes are then finally sintered. The sintering methods vary according to the type of alloy, but most types, particularly the carbide, boride, nitride and silicide alloys, are sintered in vacuum furnaces of varying design. Some of the complicated shapes have to be supported on refractory supports or saggars during sintering, particularly when bond contents are high. Oxide-base ceramals are sintered in a variety of ways, depending on the composition and type of bond material. Some are sintered under oxidising conditions, in gas-air muffles, some in molybdenum-wound hydrogen muffles, and some in high vacuum.

Another method of component production is the impregnation or infiltration method, in which a skeleton of the refractory constituent is made by cold pressing, sintering, and then machining to shape. The shaped skeleton is then re-heated in the presence of the liquid metal or alloy of the desired composition, and this liquid is drawn up into the refractory skeleton by surface tension forces, completely filling all the pores. The infiltration method is very useful in cases where the melting point of the refractory is very much higher than that of the metal or alloy phase, particularly if the metal "wets" the refractory but does not dissolve any appreciable amount of it. The electrical contact alloy tungsten-copper, in which difficulty is often experienced in obtaining theoretical densities if the metal powders are mixed, cold-pressed and sintered, is an example of this type.

The other method of producing components is the hot-pressing technique, in which heat and pressure are simultaneously applied to the powder contained in a refractory mould. The choice of mould materials is severely limited, as the mould must have high strength

at the high temperatures used, must have good thermal conductivity, must not react with the alloy, and must be easily machined. Graphite is the most common mould material, and several "washes" have been devised by which the life of the moulds is increased, and which prevent the alloy reacting with the mould. The graphite moulds are usually heated by high-frequency induction methods, or by direct resistance methods, and the pressing can be carried out in a hydrogen or other protective atmosphere, or in vacuum, if desired. Very few components are produced by this method.

Metal-Carbide Alloys

(A) Titanium Carbide Alloys

The most useful and interesting metal-refractory alloys produced in any quantity up to the present time are the metal-carbide alloys, particularly cemented titanium-carbide compositions. The high hardness and strength, good thermal conductivity, low thermal expansion and general structural stability of titanium carbide are well known, and, in addition, the fact that alloys of titanium carbide and nickel, cobalt, or iron can be sintered with a liquid phase in equilibrium with the carbide skeleton, has facilitated the production of sound, strong components, having high creep strength, high resistance to thermal shock, and, after suitable alloying, good oxidation resistance. Further, the wealth of experience acquired over many years in the production of cemented tungsten carbide alloys, some grades of which often contain large amounts of titanium carbide, tantalum carbide, etc., has been of considerable value. Alloys containing much tungsten carbide are not themselves suitable for use for moving components at elevated temperatures, owing to their high density and poor oxidation resistance. This is an unfortunate fact, as tungsten carbide-cobalt alloys have very high transverse rupture strengths, unequalled by any other metal-carbide composition of similar bond composition by volume, a fact which was repeatedly observed during the course of many war-time investigations into the possible production of tungsten-free cutting tool alloys of comparable strength, hardness and cutting performance. Titanium carbide is, on the other hand, an ideal base for high-temperature metal-carbide alloys. It has a low density, is comparatively cheap and easily prepared, and has much better oxidation resistance than tungsten carbide. The oxidation resistance can be further improved by suitable alloying, and two important series of alloys based on titanium carbide have been introduced during the last few years.

Nickel-Bonded Titanium-Chromium Carbide

One such series, on which work started in 1944, has led to the development of the Turbide alloys.² The structure of these alloys consists of a solid solution of chromium carbide in titanium carbide, bonded by a nickel-chromium, cobalt-chromium, or nickel-cobalt-chromium alloy. The alloys generally contain high proportions of nickel and/or cobalt—usually 30–50% by weight—but sometimes over the range 10–60% by weight. Work was first concentrated on alloys containing about 50% nickel, as it was intended ultimately to produce turbine blades in this type of alloy, and it was desired that the alloy should have the highest possible room temperature ductility, and high transverse rupture strength. These particular properties were required because the material was replacing a conventional

forged blading alloy which had much higher ductility, without any modification in the blade design.

The room temperature properties of these alloys vary considerably with bond content; the transverse rupture strength increases, and hardness decreases as the bond content increases over the range 10-60% by weight. It is interesting to note that Harwood³ and Graham⁴ have not observed this continuous increase in transverse rupture strength with bond content over the range 10-60% bond by weight, but that they report a maximum in the curve at about 40% nickel, above which bond content they observe a decrease in the transverse rupture strength of their Kentanium nickel-bonded titanium carbide alloys. No such maximum value is observed, however, in the Charpy impact resistance (unnotched test-pieces), and the curve given by Graham shows the impact resistance rising continuously from 4.0 in. lb. at 20% nickel to 8.2 in. lb. at 70% nickel. As would be expected, the high-temperature stress rupture properties of the Turbide alloys increase as the bond content decreases, and, consequently, a compromise has to be made between high room temperature ductility and transverse rupture strength on the one hand, and high-temperature creep strength on the other. Another important consideration is that alloys containing high proportions of bond are more liable to distort badly during sintering, owing to the large amount of liquid phase present at the sintering temperature.

These alloys are made by mixing nickel, cobalt or, in some cases, iron, with a solid solution of titanium and chromium carbides, and pressed components are sintered in vacuum at temperatures between 1,250° C. and about 1,650° C., depending on the bond content. The solid solution of titanium and chromium carbides is produced at high temperatures, and during sintering, as a result of the rapid change in solid solubility of chromium carbide (Cr_3C_2) in titanium carbide with fall in temperature, excess chromium carbide is rejected from solid solution into the liquid binder phase. On subsequent cooling, the bond freezes as a nickel- or cobalt-chromium-carbon solid solution (Fig. 1), and if the initial chromium carbide content of the carbide phase is above a certain value, free chromium carbide is also present (Fig. 2). The results of these reactions are important from two considerations. Firstly, the oxidation resistance of a solid solution of chromium carbide in titanium carbide is much higher than that of pure titanium carbide; secondly, the diffusion of chromium into the metallic bond increases not only the oxidation resistance, but also the creep resistance of this phase. The increase in creep resistance is not continuous, and a maximum is observed at the composition at which free chromium carbide is rejected from solution in the bond.

The creep strength can also be increased by decreasing the amount of bond, or by replacing part or whole of the nickel bond by cobalt. Additions of small amounts of tantalum and niobium carbides do not increase the oxidation resistance of chromium-containing alloys. The creep resistance of a medium nickel alloy, R.34, over a range of temperatures is given in Table I. In comparing these values with those of other alloys, the low density of this alloy should be borne in mind. For instance, its density is only about 75% of that of Nimonic 95. The high-temperature fatigue resistance of alloys of this series is of the same order as that of conventional forged modern high-temperature materials, and bears the usual relation to the creep strength.

The resistance of these alloys to creep under compressive stresses is interesting, in that the strain rate under compression is many times smaller than that under the same tensile stress. In this respect, these alloys differ from wrought high-temperature alloys, and this high resistance to creep in compression is due, presumably, to the high compressive strength of the carbide skeleton. Dies for hot compacting made of these materials have been used successfully for the production of hot pressings from other metal powders, operating under compacting pressures of 17 tons/sq. in. at 900° C. for long periods.⁵

Many components have been produced in these alloys, mostly by shaping in the pre-sintered state before final sintering, but also by hot pressing. These include many types of turbine rotor and stator blades, some of which required only very light lapping after sintering.

The results of one engine test using blades of these compositions have been published,⁶ in which six blades replaced the normal Nimonic 80A rotor blades, with no modification in design. These blades ran successfully for over 100 hours, consisting of several cycles of acceleration, cruising conditions, and deceleration. No failures were encountered in the blade portion, but the low ductility caused premature failure in the fir-tree type root. The radius at the bottom of the groove in the fir-tree was very small (about 0.010 in.), and the stress concentrations produced by this "notch," and by the lack of uniformity of loading along the length of the serrations, caused cracking along the grooves in the root. More recently, one aircraft engine manufacturer has developed a new type of root fixing which should overcome this difficulty.

Titanium Carbide Bonded by Chromium and Nickel and/or Cobalt

A similar group of alloys has recently been introduced in Austria, and these are called "WZ" alloys.⁷ They consist of titanium carbide bonded by chromium and nickel and/or cobalt. The similarity lies in the fact that during the sintering of such alloys, some chromium will diffuse from the liquid binder phase into the carbide skeleton, and at room temperature the alloys will consist of a solid solution of chromium carbide in titanium carbide, bonded by an alloy of nickel and/or cobalt, plus chromium. The physical properties, density, hardness, transverse rupture strength, oxidation resistance and creep resistance properties quoted for these alloys are similar to those of the Turbide alloys, for equivalent bond contents.

Alloys Based on Titanium, Tantalum and Niobium Carbides

Much work has been done in the United States of America, along slightly different lines, and the Kentanium alloys constitute the other main series of titanium carbide-based alloys.^{4,8,9} When these alloys were first introduced, the main differences between the Kentanium and Turbide alloys were the facts that the Kentanium bond contents were much lower, resulting in harder and

TABLE I.—STRESS TO FRACTURE FOR TURBIDE R.34 (tons/sq. in.)

Temperature °C.	100 hr.	300 hr.	1,000 hr.
700	24.2	22.0	19.7
750	18.9	16.4	13.8
800	14.4	11.6	8.6
870 (1,600° F.)	9.3	7.3	5.2
900	7.2	5.4	3.6
980 (1,800° F.)	4.9	3.7	2.4

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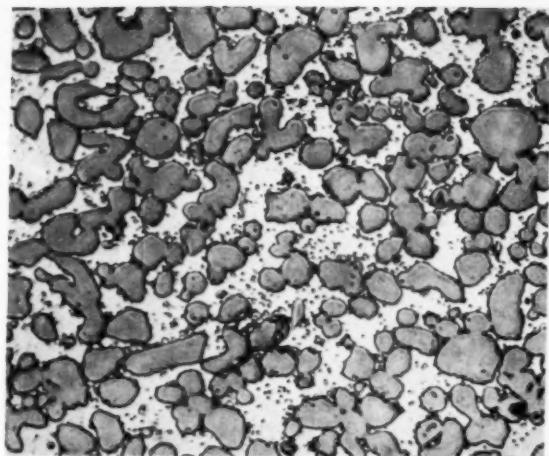


Fig. 1.—Titanium carbide phase (dark) and nickel-rich phase (light) in a high nickel-bonded alloy. Electrolytically etched in oxalic acid.

× 1500.

more brittle alloys, and that the oxidation resistance of the titanium carbide phase in the Kentanium alloys was increased considerably by additions of tantalum-niobium carbides (TaC-NbC). The first alloy of the series, K138, consisted of a straight 20% cobalt-80% titanium carbide alloy, but K138A grade was later introduced, containing the same bond content, but with additions of tantalum-niobium carbides. This range of alloys was later increased to include both nickel- and cobalt-bonded alloys containing 5-30% bond, of which the nickel alloys were considered superior.

As would be expected, additions of tantalum-niobium carbides do not increase the creep resistance of the binder phase, and the creep resistance of these comparatively low bond alloys is often lower than that of higher bond Turbide or WZ alloys in which the bond is alloyed with chromium. Reference to Table II, which summarises and compares the room and high-temperature properties of these alloys, will show that the stresses for a 100-hour life at 980° C. (1,800° F.) of Turbide R.34 (more than 40% bond), of WZ 1b (32% nickel, 8% chromium), and of WZ 1c (40% nickel, 10% chromium) are as high as, or higher than, those of alloys K138A or K151A, both of which contain 20% "un-alloyed" bond. As a consequence of these observations, the Kentanium series has been expanded during the last year,⁴ and the plain unalloyed nickel or cobalt bond has been replaced by a more complex alloy binder, to form the K16 series of alloys.

Much work has been done on the production of titanium carbide-base alloys by the infiltration method. Lidman and Hamjian¹⁰ quote interesting properties for titanium carbide and "titanium-base double carbide" skeletons impregnated with Vitallium, Nichrome, Inconel, and two types of stainless steel, but the volumes of the bond are not stated; the room temperature transverse rupture values cover a wide range, e.g. 62,000-196,000 lb./sq. in. for a pure titanium carbide skeleton impregnated with Vitallium. The oxidation resistance of the alloys made with a titanium-base double carbide skeleton (presumably containing niobium carbide or tantalum-niobium carbides) is no better than, and in most cases slightly inferior to, those made from a plain

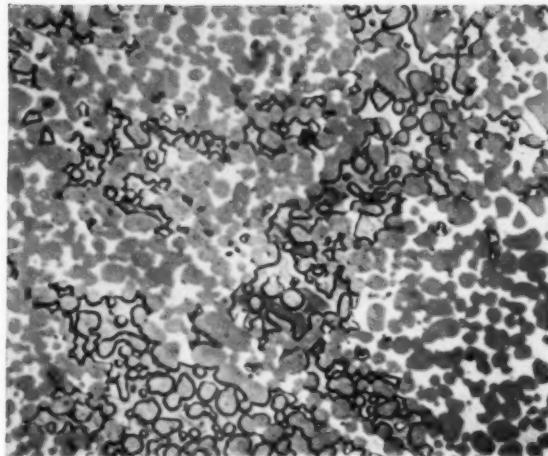


Fig. 2.—Free chromium carbide (Cr_3C_2) in a high nickel-bonded titanium carbide alloy containing chromium carbide. Electrolytically etched with 3% caustic potash.

× 1500.

titanium carbide skeleton. This is presumably because some chromium from the "superalloy" infiltrant has diffused into the titanium carbide skeleton, thus improving its oxidation resistance. Kieffer and Kolbl⁷ emphasise the usefulness of the impregnation methods in cases where only a small amount of carbide is available with which to conduct a thorough series of tests on different bonds. Nickel-chromium-base alloys were used with titanium-carbide and titanium-molybdenum carbides ($\text{TiC}-\text{Mo}_2\text{C}$) skeletons, and good hardness and room temperature transverse results were obtained. For larger scale work, Kieffer and Kolbl state that they prefer to use conventional mixing, cold pressing and sintering methods, owing to the fact that impregnated alloys often show segregation of the bond and shrinkage cavities.

Transverse rupture strengths of titanium carbide-cobalt alloys have been quoted for temperatures up to 2,400° F. (1,315° C.), and very low values are recorded at this temperature.¹⁰ Titanium carbide-base ceramics bonded with cobalt, nickel, or iron, or alloys based on these three metals, will have very little strength above such temperatures owing to the formation of the eutectic liquid phase, the transverse strength recorded at these temperatures being largely the strength of the carbide skeleton.

According to Engel,¹¹ silicon should also bond titanium carbide to a limited extent, but no results are available. Koshuba and Stavrolakis¹² quote interesting properties for titanium carbide bonded with chromium, iron, nickel-aluminium, 50-50 ferro-silicon, and Durion, but all these alloys are very brittle, and have low transverse rupture strength. Lidman and Hamjian¹⁰ describe the properties of tungsten- and molybdenum-bonded titanium carbide. No details are given of the structures of these alloys, but it would be expected that such strong carbide-forming metals would exist either in solid solution with the titanium carbide, or as free carbides, rather than as free metals.

(B) Alloys Containing Other Carbides

Little work has been reported on other carbide-based metal-refractory alloys compared with that on titanium

TABLE II.—PROPERTIES OF THE PRINCIPAL METAL-REFRACTORY ALLOYS

Alloy	Composition %	Specific Gravity	Hardness V.P.N. except where otherwise stated	Transverse Rupture		Stress-Rupture Properties				Reference
				Temperature	Strength lb./sq. in.	Temperature °C.	100 hr. lb./sq. in.	300 hr. lb./sq. in.	1,000 hr. lb./sq. in.	
Turbide R34	Medium Nickel TiC-Cr ₂ C ₃	6.15	800	R.T. [†]	150,000	800	32,300	26,000	19,300	
Turbide R45	High Nickel TiC-Cr ₂ C ₃	6.65	670	R.T.	180,000	750	31,500	28,000	23,000	
WZ 1b ..	60 TiC 32 Ni 8 Cr	6.20	1010	R.T.	190,000–210,000	800	46,000	41,400	37,000	7
WZ 1c ..	50 TiC 40 Ni 10 Cr	6.40	830	R.T.	210,000–240,000	980	14,000	11,250	8,200	3
WZ 2 ..	60 TiC 28 Co 12 Cr	6.10	1160	R.T.	156,000–177,000	980	10,300	9,100	7,750	3
WZ 12a ..	75 TiC 15 Ni 5 Co 5 Cr	6.00	1220	R.T.	148,000–163,000	980	9,500	6,800	—	3
K138A ..	80 TiC-TaC-NbC 20 Co	5.8	89.5 Ra [®] 1400	R.T.	150,000	980	11,500	—	—	3
K151A ..	80 TiC-TaC-NbC 20 Ni	5.8	89 Ra 1350	R.T.	150,000	980	11,500	9,500	7,400	9
K152B ..	70 TiC-TaC-NbC 30 Ni	5.9	85 Ra 1000	R.T.	180,000–190,000	980	5,000	3,000	—	9
K161B ..						980	15,500	—	—	4
K162B ..						1095	5,500	—	—	4
K162C ..						980	12,500	—	—	4
K163B ..						980	13,000	—	—	4
ZrC-Nb ..	87.5 ZrC 12.5 Nb	6.29		1,093° C. 1,315° C.	25,200 15,900					13
B ₄ C-Fe ..	64 B ₄ C 36 Fe	3.24		1,093° C. 1,315° C. 1,426° C.	25,200 28,600 23,400					16
CrB-Ni ..	88 CrB 15 Ni	5.9	87.4 Ra	R.T.	193,000	816 872	4,600 2,600	4,200 2,200	3,500 1,700	22
Metamic LT-1 ..	70 Cr 30 Al ₂ O ₃	6.0	35 Ra							25
Al ₂ O ₃ -Cr ..	70 Al ₂ O ₃ 30 Cr	4.6– 4.65	1100– 1200			980 1043 1203	16,200 13,500 12,000			26
Nimonic 90	Forged Ni-Co-Cr Heat-treated alloy	8.27	250– 350			815	28,000	23,500	18,000	27
Nimonic 95	Forged Ni-Co-Cr Heat-treated alloy	8.3	approx.			815 870 900 925	31,400 20,200 14,600 11,200			27

* Ra signifies Rockwell "A" scale and Rc Rockwell "C" scale.

† R.T. signifies Room Temperature.

carbide-based alloys. Alloys containing large amounts of tantalum, niobium or vanadium carbides are not used for such applications, principally because of expense. Zirconium carbide should be a useful refractory component, being similar in many respects to titanium carbide, but bonding difficulties may be the reason for the fact that few alloys of zirconium carbide (ZrC) are being studied. Its oxidation resistance is inferior to that of titanium carbide at temperatures up to about 1,800° C., but, presumably, it could be improved by suitable alloying as has been done with titanium carbide. One alloy reported, a niobium-bonded zirconium carbide alloy made by hot pressing,¹³ is another example of the use of a strong carbide-forming metal as a bond. During the hot pressing, the niobium carburised at the expense of the zirconium carbide, and free zirconium was identified in the final structure.

Molybdenum carbide is not used as an alloy base owing to its poor oxidation resistance, although it is sometimes used as small additions to titanium carbide-base alloys.

Chromium carbide-nickel alloys¹⁴ have attracted considerable attention during the last few years, as a new series of cemented carbide material for room temperature applications such as gauges, wear-resistant parts, and corrosion-resistant components. High-temperature properties have not yet been published, but the nickel-chromium-carbon bond should have good creep and oxidation resistance. The transverse rupture strength of these alloys is low, but it could possibly be improved by suitable alloy additions. Silicon carbide has well-known refractory properties, but little work has been reported on alloys containing this carbide. Silicon carbide (SiC) has been bonded with metallic silicon by infiltration and some evidence of the "wetting" of silicon carbide by metallic silicon has been observed.¹⁵ One of these compositions retains its tensile properties at elevated temperatures, e.g. 65,000 lb./sq. in. at room temperature, and 60,000 lb./sq. in. at 1,320° C. (2,400° F.), and has a transverse rupture strength of 33,000 lb./sq. in. at 1,320° C. (2,400° F.). Iron-bonded

boron carbide (B_4C),¹⁶ and beryllium-beryllium carbide alloys¹² have been produced. The effect of high temperature on the transverse rupture strength of a cold pressed and sintered alloy containing 64% boron carbide and 36% iron is interesting, as the rupture strength is reported as 32,600 lb./sq. in. at 1,600° F. (872° C.), and 23,400 lb./sq. in. at 2,600° F. (1,427° C.). The density is only 3.24.

Complex carbides are seldom used as alloy bases, but Goldschmidt¹⁷ describes a process for the production of "cutting tools or heat-resistant alloys," in which complex carbides are extracted from suitably heat-treated steels, by electrolytic or other methods, and these carbides are then bonded by a variety of metals including cobalt, nickel, tungsten, molybdenum and titanium. Specific carbides mentioned are the iron-tungsten double carbide, analogous to the brittle cobalt-tungsten double carbide which is usually regarded as detrimental to the properties of cemented tungsten carbide alloys, and the corresponding iron-molybdenum double carbide.

Metal-Nitride Alloys

The interstitial nitrides have properties and structures analogous with the carbides, and in many cases form extensive or complete series of solid solutions with their respective carbide, e.g. TiN - TiC , ZrN - ZrC . The oxidation product of carbide- and other refractory-base materials contains nitrides and oxides, and, consequently, the refractory properties of nitrides are important. No metal-nitride alloys of importance have been reported, although the properties of an alloy consisting of 50% magnesia, 30% titanium nitride, and 20% nickel oxide,¹⁸ during the sintering of which some nickel oxide is reduced to nickel, have been reported. The transverse rupture strength of this alloy increases from 20,000 lb./sq. in. at 1,500° F. (816° C.) to 31,000 lb./sq. in. at 2,400° C. (1,316° C.). This increase is alleged to be due to the oxidation of the nitride. Titanium forms two nitrides, TiN and Ti_2N ¹⁹ and of these, the former deserves much more study, possibly as a major addition to titanium carbide-base alloys.

Metal-Boride Alloys

The interstitial borides are at the moment of much less technical importance than the carbides, but work is now in progress which might well lead to their more widespread use. Kieffer and Benesovsky²⁰ have summarised their properties and methods of production, while Keissling²¹ has published many results of determinations of crystal structure, and of metal-boron systems. The interesting features of the metal-boron systems are the very high number of borides formed, e.g. the metals chromium, molybdenum and tungsten each form at least three. The existence of so many borides might lead to the production of very useful metal-boride alloys, but few details have been made available to date.

Sindeband²² has published details concerning alloys of chromium boride bonded with nickel, nickel-copper, nickel-chromium, and cobalt. The nickel alloys were most promising, and a hot pressed alloy containing 15% nickel had a room temperature transverse rupture strength of 123,000 lb./sq. in. The alloy had comparatively low stress-rupture properties, but good oxidation resistance up to 950° C. (1,742° F.). A liquid phase was observed at about 1,900° F. (1,038° C.), reputed to be

due to the formation of a less refractory nickel boride. The formation of this liquid phase limits the use of the alloy to about 1,750° F. (954° C.), and the author considers that the properties at 1,500° F. (816° C.) and 1,600° F. (872° C.) are also affected. Bonded zirconium boride alloys are also reported, but few properties have been quoted.

Metal-Silicide Alloys

Metal silicides have generally high melting points and good oxidation resistance, but little is known of the properties of bonded silicides. Certain of them are of considerable interest, particularly molybdenum disilicide ($MoSi_2$), which has a melting point of 2,030° C. ± 50° C.,²⁰ good oxidation resistance, and can be bonded with silver.²³ This silicide is also of interest because molybdenum and its alloys can be protected from oxidising atmospheres by a siliconising process, which produces a surface layer containing a series of molybdenum silicides, including $MoSi_2$.

Metal-Oxide Alloys

Metallic oxides are generally recognised as belonging to the group of materials known as ceramics, as they have low thermal and electrical conductivities and general "non-metallic" behaviour. Some alloys formed between these oxides and metals are different from those between metals and carbides, nitrides, borides and silicides, in that their resistance to thermal shock (in the case of alloys containing small proportions of metal) is generally poor, and their transverse rupture strength is usually low. For obvious reasons their oxidation resistance is very good, and, consequently, metal-oxide alloys are sometimes regarded as the class of metal-refractory combinations which will eventually withstand the highest service conditions.

Mixtures of iron and alumina were first used many years ago, particularly in Germany, where several tests on alumina-iron turbine rotor and stator blades were conducted.²⁴ Alumina-iron alloys were found to have very good thermal shock properties, and it was found that the electrical conductivity of alumina-iron alloys increases very rapidly as the iron content rises to about 50% by weight, and above, presumably indicating the formation of an interconnected skeleton of the iron-rich phase. The alloys are usually made by mixing alumina and fine iron powder (carbonyl iron), cold pressing and sintering, usually in an inert or reducing atmosphere. Sintering temperatures are usually high, contraction occurring in the temperature range 2,560°-2,870° F. (1,405°-1,577° C.), i.e. above the melting point of pure iron.

More recently, chromium-alumina alloys have been investigated and have given promising results. The compositions studied cover a wider range of chromium contents, and properties for alloys containing 30% chromium and 70% chromium are reported. The alloy 70% chromium-30% alumina (LT-1) has very interesting properties.²⁵ The structure consists of free chromium bonded by a spinel-type bond to the alumina. During oxidation, the free chromium at the surface forms a protective film of green chromium oxide. The electrical conductivity is many times that of pure alumina, and the thermal shock resistance is said to increase after the first few thermal shock cycles. The low bond 30% chromium alloy also sinters very well, to produce alloys containing very little porosity, having a density of 4.60-4.65 g./cu. cm., compared with the theoretical

4.68-4.72%; the hardness is high at 1,100-1,200 V.P.N. The resistance to oxidation is excellent up to 2,750° F. (1,510° C.) and this alloy also shows an increase in thermal shock resistance after the first few temperature cycles.

Other oxide-base ceramals include the magnesia-titanium nitride-nickel oxide alloy already discussed, and a beryllium-beryllia hot pressed alloy,¹² but few results are available.

Summary

The properties of the principal metal-refractory alloys are given in Table II, in comparison with those of Nimonic 90²⁷ which is probably the best wrought blading alloy in large scale use throughout the world at the moment. The stress-rupture properties of some of these alloys compare very favourably with those of Nimonic 90, without making any allowance for differences in specific gravity, which will be significant in rotor blading applications. At the present time, the most important of all these alloys is the group of alloys based on titanium carbide. More is known about these alloys, concerning the production technique, room temperature and high-temperature properties, and the behaviour under test in a gas turbine, than about any other type of metal-refractory alloy. Even these alloys, however, are not yet ready to replace conventional forged blading alloys in gas turbine rotors, and the engine tests carried out to date represent only the first stage in the introduction of new high-temperature materials. Further progress in the use of these alloys now depends to a large extent on the co-operation of engine designers, who must determine the best methods of attaching the blades to the turbine discs, when using materials having lower room temperature ductility than the forged alloys. The simplification of the complicated shapes would also assist the development of the production of turbine blades in these materials, as most of the shaping must be done before the final sintering stage, as the sintered blades are very hard, and can only be machined by expensive diamond grinding, or by the recently introduced electro-machining technique.

The possibility of producing sintered blades cheaper than precision forged blades is extremely doubtful, and the sintered blades can therefore compete on the grounds of their better high-temperature properties only. The titanium carbide-base alloys can probably be used at higher temperatures than Nimonic blades, but this increase in temperature is limited, as the alloys will have very little tensile strength at temperatures of 1,300° C. and above, owing to the formation of liquid phase. If any temperatures higher than these are required, presumably the chromium-alumina "cermet" will be used. Such blade temperatures, which are excessive at the present day, raise other problems of design and construction which will be much more difficult to solve than merely the choice of suitable blading alloys. Obvious difficulties are the choice of material and construction of flame tubes, combustion chambers and compressor blading and design problems. In other words, the titanium carbide-base alloys will probably be the most important type of metal-refractory alloys for some years to come. By that time, much more will be known about the production and properties of other alloys, particularly the alumina alloys, and the experience which should have been gained in the use of other hard non-ductile materials will be of extreme value.

The cooling of blades by the "sweat-cooling" technique is a method with which powder metallurgy methods are usually associated, but the production of porous metal-refractory blades having controlled interconnected porosity would seem to be a very difficult problem. In any case, if sweat-cooling methods are used, the blade temperatures would be reduced and the use of metal-refractory alloys might become unnecessary.

Acknowledgments

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Electrical Engineers' Exhibition

The Association of Supervising Electrical Engineers is to hold its Third Electrical Engineers' Exhibition at Earls Court, London, from 16th-20th March, 1954. The success of the last exhibition has led to so many requests for a longer show that the exhibition next March will extend over five days instead of four, and will be open to the trade only. Switchgear will be the principal feature and two awards will again be made to the manufacturers exhibiting the most outstanding 1954 domestic labour-saving and industrial devices. Electrical Research Association, British Electricity Authority, British Electrical and Allied Manufacturers Association, British Electrical Development Association, Cable Makers Association, and other organisations serving the industry are again taking part.

Change of Address

THE MULLARD PUBLICITY AND PRESS DEPARTMENTS now occupy new offices at 1, Gerrard Place, Shaftesbury Avenue, W.C.2 (Tel.: GERrard 9941 to 9945). All correspondence should continue to be addressed to Century House, Shaftesbury Avenue, W.C.2.

The Cold Pressure Welding of Metals

By J. E. Hughes, B.Sc., A.R.S.M., Ph.D., D.I.C.*

Using a method of cold welding cylindrical rod specimens in which the deformation may be confined to the interfacial zone, the weldabilities of aluminium, copper and iron have been studied. Results on combination welding suggest that the factors which determine the weldability of a metal play a significant part when the metal is welded to another. The metal-oxide relative hardness theory is discussed.

PRESSURE welding may be defined as the process of joining metals under conditions such that no liquid phase is produced at, or introduced into, the weld at any stage. Cold welding is the name given to pressure welding when no heat is supplied to the system and all operations are carried out at room temperature. It is only within recent years that attention has been paid to cold welding and the possible uses which it may have for fabrication purposes, but a considerable amount of information has already accumulated.^{1, 2, 3}

Cold welding utilises the natural surface forces of metals and, therefore, contaminant films (oxides, grease, adsorbed films, etc.), have to be removed, and sufficient deformation of the interfacial zone must be produced to bring large areas of the contiguous surfaces into intimate contact. In general, the bonding forces between the atoms in a given metal are higher at low temperatures, but welding is more difficult because the low plasticity necessitates high deformation pressures. In addition, other factors, such as the low diffusion rates of contaminants, lack of recrystallisation, etc., affect the welding.

Most of the work on cold welding has been concerned with aluminium and its alloys because, of the commercial metals, it is the easiest to weld, requiring small deformations and pressures. An added impetus to this work has been provided by the difficulties of soldering and fusion welding. However, it has now been demonstrated that a large number of metals and alloys can be cold welded, although in many cases very large deformations are required to produce high strength joints.

Almost without exception, work on cold welding has been confined to the investigation of sheet specimens, the method used being shown in Fig. 1a. The parameter used to define the amount of deformation of the weld zone is the "figure of merit."³ This is defined as the "maximum percentage of the double thickness of the sheets which can remain when a sound weld is produced."

Using this concept, it has been shown that metals vary considerably in their figures of merit, and work is now progressing to account for the results obtained.

Theoretical Considerations

It may be stated that no completely satisfactory theory is available to explain why one metal should be more easily welded than another. The cohesion between the metal and its oxide is an important consideration, and factors which influence this are likely to affect the weldability. Whitehead⁴ has suggested that the relative crystalline orientation and degree of misfit of the lattices of the metal and oxide might influence the oxide adherence. In the case of copper, the oxide has probably a similar orientation to that of the underlying metal,⁵ whereas there appears to be no evidence of such a similarity in the case of aluminium.⁶ An important factor may be the relative hardness of metal and oxide. This would account satisfactorily for the greater ease of welding of aluminium than of copper. On first consideration it would appear that aluminium, with its hard, tenacious oxide, would be more difficult to weld than copper. However, the reverse is the case. It is considered that, as the oxide film on aluminium is much harder than the metal beneath, it will be easily ruptured during deformation, so producing intimate metallic contact and consequent welding. Analogous with this is the ease with which a layer of ice on soft mud or water is broken when pressure is applied.

Indium has been said to cold weld immediately on contact, and this has been ascribed to the absence of an oxide film. Silver possesses an oxide film, and high deformations are required for welding, but if the film is first dissociated by heating to 700° C., welding can take place with much smaller deformations.¹ It has been reported⁷ that lead requires a deformation of 84%, but it has since been shown that 20% is sufficient if welding is carried out immediately after surface preparation.¹

Hudson⁸ has suggested that cold weldability may be

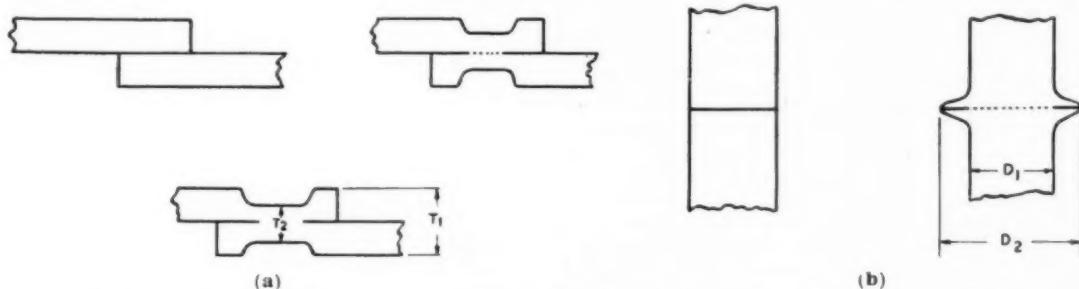


Fig. 1.—Pressure welding of (a) sheet specimens and (b) cylindrical rod specimens.

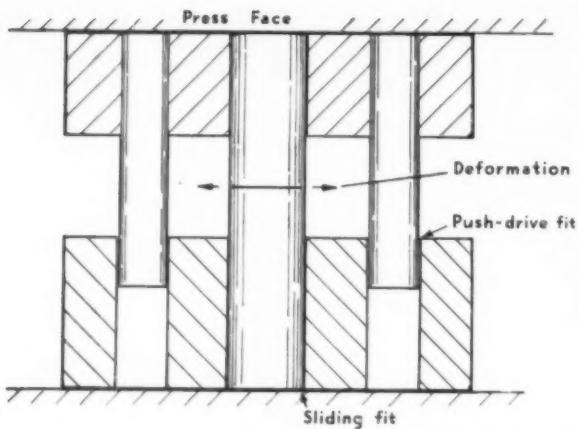


Fig. 2.—Jig for cold welding cylindrical rod specimens.

a function of the ability to recrystallise under the influence of cold work. It is pointed out that there is evidence in powder metallurgy that the sintering temperature falls as the grain size is decreased, and an analogy is drawn between this and massive metal surfaces when deformed under the welding pressure. This suggests that a large number of crystallites are formed as a result of the cold work, and that grain growth can occur at temperatures sufficiently low to obviate the necessity for external heating. This theory appears unsatisfactory in view of the cold welding of dissimilar metals, and particularly as it has been demonstrated⁸ that metals which show no solid solubility in each other can be cold welded together.

The above theories may be most easily examined by investigating a number of systems of known properties, including the hardness of oxide, hardness of metal, thickness and time of formation of oxide, etc. In addition, a significant quantity which should be determined is the deformation required to produce just sufficient adhesion to hold the surfaces in contact, as well as the figure of merit required for high strength joints. A useful test for the theories would be the welding of dissimilar metal combinations, whereby, for instance, the relative ease of welding aluminium to itself and to other metals could be compared.

Cylindrical Rod Test Pieces

Although there are advantages in the testing of sheets for cold weldability, such as the small amount of material required (an important consideration for rare or precious metals), the method is open to a number of objections. The metal is drawn down round the periphery of the weld tools and consequently weakened, so that when mechanical tests are made, a tendency exists for the weld to break at this point, with the erroneous inference that the weld is as strong as the parent metal. In addition, it is difficult to test adequately welded sheet specimens. The normal method is to pull the specimens and bring the weld interface under shear. However, a bending moment is soon introduced, with the consequence that the property being measured is somewhat nebulous. It was anticipated that difficulties would be encountered when welding dissimilar metal specimens. If the metals had widely differing properties, such as hardness, the softer metal would tend to flow under a given pressure

without adequate attenuation of the harder surface, and contaminant films would tend not to be broken. The harder of the two metals would also be forced into the softer, and the interface would not lie in the same plane as the junction between the sheets in the unwelded parts.

It was concluded that, to supplement information obtained from sheet specimens, pressure welding tests on cylindrical rod specimens would be of considerable interest. One method available is to take two cylinders of the metal under test and apply an axial load to compress the specimens. A limit is imposed on the length-diameter ratio to prevent buckling under pressure. The obvious requirement is to confine the deformation to the weld zone and leave the major part of the rod length undeformed (Fig. 1b).

An apparatus to do this is shown in Fig. 2. The jig is of heat treated En26 steel and carries the specimens in the guides as shown, the diameters having been machined to allow a sliding fit. At the beginning of the welding the top platen of the jig projects above the top of the upper specimen, but does not slide downward as the guide pins are made a push-drive fit. When the

TABLE I.—RESULTS OF COLD PRESSURE WELDING EXPERIMENTS.

Metal	Deformation %	U.T.S. tons/sq.in.	Elongation %	Remarks
Aluminium to Aluminium	33			No weld
	42			Welded—low strength
	51	3.1	Negligible	Fracture at interface
	63	4.8	25.0	Ductile fracture
	71	4.9	34.0	Ductile fracture
Copper to Copper	88	4.7	43.0	Heated to 500° C. and cooled
	40			No weld
	58			Slight adhesion
	70	10.8	1.4	Fracture at interface
	72	16.0	2.4	Fracture at interface
Iron to Iron	84	15.7	21.3	Ductile fracture
	74			No weld
	75			Limited adhesion
	84			Little strength
	90			Increased strength—broke on machining
Iron to Aluminium	47			No weld
	56			Adhesion—low strength
	66			Broke on machining
	71	5.7	Negligible	
	81	6.4	Negligible	
	81	6.1	Negligible	Fracture through aluminium
	81	6.8	Negligible	200° C. for 30 mins.
	81	10.8	1.1	300° C. for 30 mins.
	81	11.1	2.0	400° C. for 30 mins.
	81	Brittle		500° C. for 30 mins.
Iron to Copper	58			600° C. for 30 mins.
	63			No weld
	67			Just adhered
	84			Broke on machining
	84			Broke on machining
Aluminium to Copper	33			No weld
	37			Adhered
	53			Broken on machining
	77	10.3	Negligible	Fracture in aluminium
	85	11.4	Negligible	Fracture in aluminium

pressure is applied, the top platen is forced down until its upper face is in line with the upper surface of the top specimen. As the top platen descends further, the cylindrical specimens come under pressure and deform between the upper and lower platens. The pressure is increased until the required deformation has been obtained. An ejector device is used to extract the specimens after welding.

Three welded test pieces of aluminium are shown in Fig. 3. They are 0.7 in. diameter and 1½ and 2 in. long. The first specimen had 88% deformation, the second 63% and the third 33% (which, incidentally, did not weld). The deformation values reported in describing these experiments are based on the spread of the interface, being equal to

$$\frac{\text{Final Area} - \text{Initial Area}}{\text{Final Area}} \times 100\%$$

The dimensions of the welded specimens allowed suitable test pieces to be prepared for tension, shear and bending tests. For

the welding of dissimilar metal combinations, such as aluminium to copper, either one rod could be made of copper and the other of aluminium, or a sheet of aluminium could be placed between two copper rods. The latter method proved the more useful, because of the more even deformation and the advantages of mechanically testing such welds.

Results

The metals examined were aluminium, copper and iron. These were welded both to themselves and to each other.

Aluminium to Aluminium

The metal was of commercial purity. The welding surfaces were prepared by machining, followed by scratch brushing and degreasing. The minimum deformation for adhesion, that is, that deformation which was just sufficient to hold the surfaces in contact, was about 35%. This had to be increased to 60% before the welds attained strengths sufficient to fracture away from the interface in the tensile test. These values were not significantly altered by using high purity aluminium.

The tensile test pieces were machined with the interface at the centre of the parallel portion, and it was observed that, due to the cold working of the interfacial zone, fracture occurred with necking in the normal manner away from the weld plane, in the softer unworked metal, unless the strength of the weld was sufficiently low for fracture to occur through the interfacial plane. The influence of this cold-worked zone meant that the elongation percentage values could not be compared directly with those of the parent metals.

Table I gives details of the results obtained, and refers to the as-welded conditions. Heating the welds to soften the interfacial zones did not affect the U.T.S. value, but the elongation percentage increased appreciably, and fracture still occurred away from the weld plane. This indicates that heat treatments in aluminium pressure welds do not reduce the strength, due to gas rejection, a possibility noted by Tylecote.¹

Copper to Copper

Tough pitch electrolytic copper was used for these

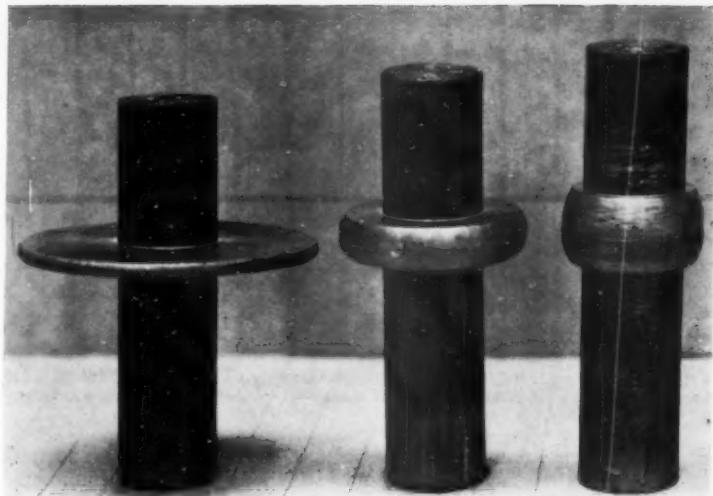


Fig. 3. — Cold welded specimens of aluminium (approximately 1/2 normal size).

experiments. The method of surface preparation was to machine the faces, polish to 000 emery, etch in nitric acid for 5 seconds, wash in alcohol, dry, scratch brush and degrease. The minimum deformation for adhesion was 55%, while about 80% was required for high strength joints. Results of tensile tests are given in the table, and it is seen that the 72% deformation weld gave approximately the same ultimate tensile strength value as that made with 84%, even though fractures occurred at the weld plane in the former but in a necked manner away from the weld plane in the latter. The elongation percentage values, however, demonstrate the differences in ductility of these types; this illustrates an effect which has often been noted—that the U.T.S. is a poor criterion of weld quality, while the elongation percentage and reduction in area percentage are considerably more sensitive to weld plane weakness.

Iron to Iron

At the time of carrying out this research, it was not known whether or not iron could be welded at room temperature. The material was commercial quality Armeo iron, and the surfaces were prepared by grinding and degreasing. It was found that such surfaces did adhere, but the minimum deformation to produce adhesion was in the range 75–80%. The deformation necessary for high joint strength could not be found, as the pressure available was insufficient for the purpose. Although 90% deformation was incapable of producing strength sufficient to allow machining into tensile specimens, the adhesion was considerable.

Iron to Aluminium

For these experiments, the aluminium was used as 1 in. thick material sandwiched between two iron cylinders, the methods of surface preparation being as before. The minimum deformation for adhesion was about 50%, high strength joints being produced by 75% deformation. All fractures in the tensile test were of the brittle type, with negligible elongation or necking. This was due to the stressing conditions of the aluminium, it being constrained from deforming by the iron. However, the quality of the weld could be assessed, as fracture in low strength welds occurred at or near one

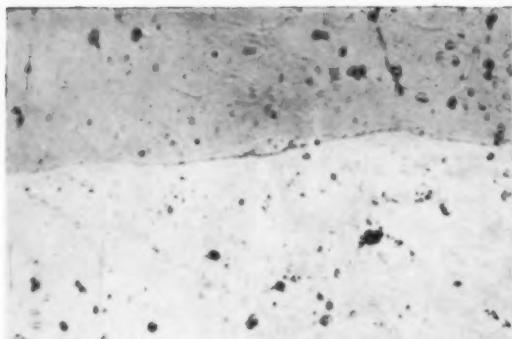


Fig. 4.—Photomicrograph of section through cold weld made between iron and aluminium with 79% deformation. Unetched. $\times 500$

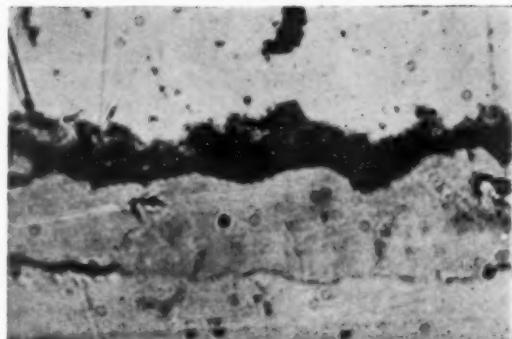


Fig. 5.—Photomicrograph of section through cold weld made between iron and aluminium with 81% deformation: reheated at 600°C. Unetched. $\times 500$

of the iron-aluminium interfaces, while in welds of high strength fracture occurred through the body of the aluminium. Heat treatment in the range 200–600°C. of iron-aluminium welds, based on 81% deformation, indicated that, for a 30-minutes heating period, no effect on weld strength resulted below 300°C. Heating at 400–500°C. increased the U.T.S. value and slightly increased the ductility, while at 600°C. the weld became embrittled, due to the formation of intermetallic zones at the interfaces, as a result of diffusion. These results support the findings of Cooke and Levy⁹ on the hot pressure welding of iron to aluminium.

Fig. 4 is a photomicrograph of an iron-aluminium cold weld, with complete elimination of porosity and oxide dispersal. Such a weld would fail in the tensile test by fracture through the aluminium. Fig. 5 shows the presence of embrittling iron-aluminium phases produced by heat treatment.

Iron to Copper

The minimum deformation for adhesion was 60%. No high strength joints were made, because of pressure limitations, but deformation in excess of 85% was necessary.

Aluminium to Copper

The cylindrical test pieces were copper, and the inserted sheet aluminium. The minimum deformation for adhesion was 35%, while 70% was required for making welds of sufficient strength to fracture through the aluminium and not at one of the aluminium-copper interfaces. In all cases, fracture occurred without appreciable elongation of the test pieces.

Discussion of Results

An examination of the values for the minimum deformation required for adhesion and to produce strong joints (Table II) for the metals and combinations under discussion, reveals the following points.

(a) Aluminium is easily welded and, in combination with other metals, is able to reduce considerably the deformation required for a given weld strength. The theory based on the relative hardness of oxide(s) and metal(s) appears satisfactory in explaining these results. Table III gives values for the hardness of the metals and oxides involved. Values for oxide are given on Moh's scale, and relate to the bulk material and not surface films: these are taken as typical in the absence of information on surface film properties.

The minimum deformation for adhesion for copper-aluminium is the same as for aluminium-aluminium. At each copper-aluminium interface there are four layers present, that is, aluminium (2–2.5), aluminium oxide (9.0), copper oxide (3.5–4) and copper (3.0). The alumina film is thus still much harder than the other materials involved, and will rupture quite easily when the metals deform under pressure. The result is as though the alumina film is able to assist in the rupture of the copper oxide film. The effect of the alumina is also seen in the reduction of deformation necessary for high strength joints in copper.

(b) The introduction of the aluminium has had a parallel effect in iron welds, bearing in mind the greater hardness of the metal and oxide in this case. The minimum deformation for adhesion is reduced from 75–80% down to 50%, and high strength joints can be obtained with the pressure available.

(c) The iron-copper combinations are difficult to weld, although the increased ductility of the copper slightly reduces the deformation required for iron to iron. However, high strength joints cannot be obtained with the pressure available.

A significant point which emerges from the results is the high strength which can be obtained from the interfacial bond of dissimilar metals. When an iron-aluminium weld fails by fracture through the aluminium component, the cohesive strength of the interface is greater than that of the severely cold worked aluminium. Even in low strength welds, and in welds which have not adhered in bulk, there is considerable pick-up of the softer metal on the harder—particularly obvious in the

TABLE II.—MINIMUM DEFORMATION NECESSARY FOR ADHESION AND FOR STRONG JOINTS.

Metal(s)	Minimum Deformation % for Adhesion	Minimum Deformation % for Fracture away from Interface
Fe-Fe	75–80	>90
Al-Al	35	60
Cu-Cu	55	80
Fe-Al	50	75
Fe-Cu	60	>85
Cu-Al	35	70

TABLE III.—HARDNESS OF THE METALS AND OXIDES CONCERNED.

Material	Hardness (Moh's Scale)
Al	2.2–5
Al ₂ O ₃	9.0
Cu	3.0
Cu ₂ O	3.5–4.0
Fe	4
Fe ₂ O ₃	6

aluminium-copper welds because of colour differences. The fracture has therefore mainly occurred a short distance inside the softer metal. In the region between pick-up and general adhesion, it is probable that elastic recovery has been sufficient to cause breakdown of the joints. Good adhesion occurs between the two surfaces only when the welded junctions are sufficient in number.

It is pertinent at this stage to note that the method used, of welding cylindrical specimens, with the deformation confined to a narrow interfacial zone, may have uses other than investigating the cold welding of metals and combinations. For instance, it would lend itself to the preparations of diffusion specimens. If the diffusion of *A* in *B* (or a radio-active isotope of *B* in *B*) were being investigated, a strip of *A* could be cold welded between two cylinders of *B* or a cylinder of *B* could be welded to a cylinder of *A*. One advantage of this method would be that diffusion specimens would not need elevated temperatures for preparation and, therefore, the amount of diffusion occurring on preparation would be insignificant.

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Conclusions

A method of cold welding cylindrical specimens, in which deformation may be confined to a narrow zone at the interface, has been utilised to study the welding of iron, copper and aluminium and combinations of these metals. Results suggest that the factors which determine the ease of welding of a particular metal also play a significant part when the metal is welded to another. A theory based on relative hardness of metal and oxide provides an adequate explanation of the observed results. Considerable pick-up of one metal on the other, even before complete adhesion, illustrates the high strength of the bond. Results of exploratory tests indicate that weld strength and ductility can be usefully increased by heat treatment, providing brittle phases are not produced.

Acknowledgments

The author wishes to thank Dr. M. S. Fisher for help and advice, and Professor C. W. Dannatt for the use of laboratory facilities.

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City and Guilds' Insignia Award

THE Insignia Award in Technology was established by the Council of the City and Guilds of London Institute some twelve months ago to provide a high qualification for persons in industry whose initial training was based primarily upon practical experience combined with theoretical study and who, having gained appropriate City and Guilds' certificates as craftsmen or technicians, have now advanced in their industry by a combination of progressive experience and further study.

A candidate upon whom "The City and Guilds of London Institute's Insignia Award in Technology" is conferred will receive a Warrant specifying thereon the section of industry and the branch of technology in which his knowledge and skill are recognised, and under the Institute's Royal Charter he will be authorised to use the insignia letters "C.G.I.A."

The Regulations state that a candidate must be at least thirty years of age and be sponsored by three referees of appropriate standing. He must have been apprenticed or otherwise suitably trained in his industry, he must hold relevant Full Technological Certificates of the Institute, and have had a minimum of seven years' progressive experience in his industry. He is then required to write a thesis of some 10,000 to 20,000 words on a technological topic associated with his work, which must show that the candidate is acquainted with the current practice and problems of his industry at "technological" level. Finally, if this thesis is reported upon satisfactorily by the Assessor appointed by the Institute, the candidate is required to appear for interview before an *ad hoc* Panel nominated by the Institute for this purpose. The Insignia Award Committee, on the recommendation of the Panel, will then make its final decision as to the Award, subject to ratification by the Council of the Institute. There is no special time-table involved in the scheme, and each candidate is considered as a separate and self-contained "case," and great care

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It will be appreciated that in one year the stage has not yet been reached when a first list of recipients of the Award can be issued, but candidates' theses are now being received and considered. There is every indication that the wide range of topics and the standard and level of treatment dealt with in the candidates' theses carry great possibilities as to contributions to technological practice and discussion. The Insignia Award Committee is considering a scheme whereby a précis of those written by successful candidates may be made available to all who are interested and, in exceptional cases, where guidance may be given to ensure full publication.

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Grinding Hazards

By W. R. Hardwick, A.Met., A.I.M.

Arthur Balfour & Co., Ltd., Sheffield.

The grinding of tools is accompanied by local heat generation, and unless the operation is carried out correctly it can result in softening and cracking of the surface of the tool. These effects may be aggravated by incorrect heat treatment prior to grinding.

MOST of the cutting tools in use today are ground in some way after heat treatment, an operation which can vary from simple hand grinding to elaborate form grinding in specialised machines. It has been our general experience that trouble due to the formation of grinding cracks is almost non-existent in the case of hand ground tools, and the same applies to fine form and cutter grinding, the largest number of complaints arising from surface ground tools such as dies, punches, etc. In the majority of cases, the customer is not aware that a tool failing in service did so because of the presence of grinding cracks, and the investigator is usually informed that such cracks could not have been present as "only 5 thou was taken off." Fig. 1 shows a moulding die which broke in use, one half of which has been etched to show the serious cracking present.

There is little doubt that the removal of even 0.002 in. from the surface of a very hard tool, with an unsuitable wheel or speed, can produce cracking, and if overheating in hardening has occurred, or tempering been omitted, the chances of producing a sound, ground tool are very greatly reduced. Apart from the production of surface cracks, it is not unlikely that the ground surface will be slightly softened as a result of the tempering effect: the phenomena of cracking, lifting out of the surface, and softening are bound up with each other, all being caused by the local generation of heat.

Taps and Hobs

The production of ground thread taps and hobs involves the removal of relatively large amounts of metal to form the teeth of the tool concerned, and in many ways the difficulties in grinding are likely to be more severe than in other forms of grinding, for several reasons:

- (1) The tool is made from an abrasion resistant material such as high speed steel or high-carbon high-chrome steel.
- (2) The tool blank has to be thread ground at a hardness in the range 63-66 Rockwell "C."
- (3) The tooth tips, being of small section, are soon heated to a temperature high enough to produce softening, and the resultant damage is

not easily detected. The use of coolants is no guarantee that local heating can be prevented.

In manufacturing high speed steel ground-thread hobs, it is general practice to machine the tool in the usual manner from annealed bar stock, complete with the requisite gashes. The portion of the tool which will ultimately contain the teeth is left untouched, except for removing bar skin and machining the necessary relief.

The tool blank is then hardened and tempered to a hardness of 63-66 Rockwell "C" following which the teeth are generated by grinding. This grinding operation, carried out on the hard blank to some considerable depth, has to be done with care to avoid softening the cutting edges of the tool, and even under the best conditions it is often found that the temperature attained on the tooth tips during grinding will reduce the hardness value by one or two points on the Rockwell "C" scale.

The slight drop in hardness may be of no serious consequence, although the tooth tips will tend to rub off and the exact form be lost, but if the grinding is done under unsuitable conditions the damage, although not immediately apparent, can be quite a serious problem to the tool maker and user. Premature wear takes place on the tips of the teeth due to the local softening

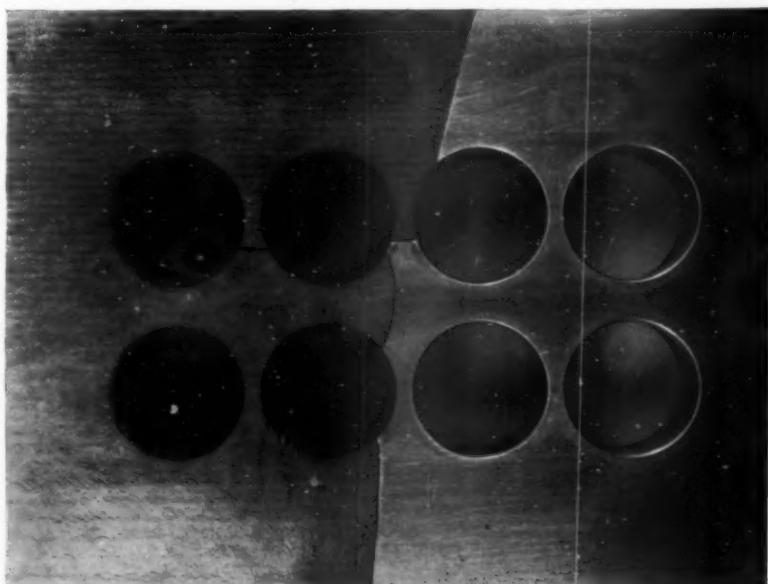


Fig. 1.—Moulding die which broke in service. Left hand piece has been etched to show serious cracking.

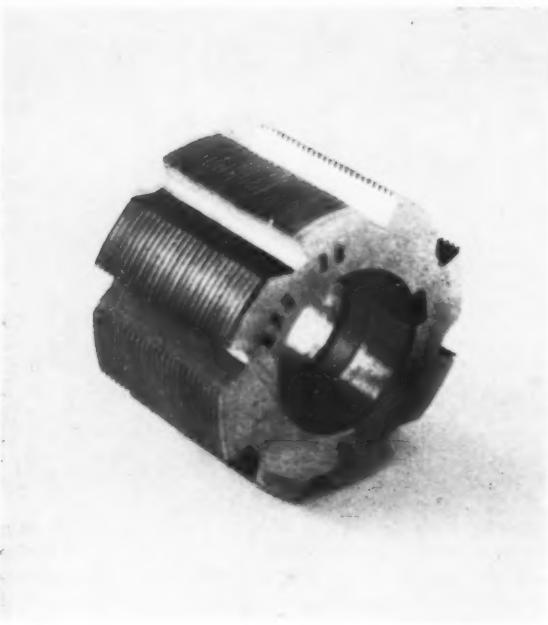


Fig. 2.—Typical hob found to be cracked after grinding the teeth.

effects, and fine cracks become visible on tools which have been badly ground. Fig. 2 shows a typical hob which was found to be cracked after grinding or forming the teeth, and Fig. 3 shows a similar tool where the cracked portion lifted out.

In the case of the tools referred to above (Figs. 2 and 3) it was found, by careful testing with a small file, that the tooth tips could be filed off, this being apparent whether cracking had occurred or not. After removing a section of the affected teeth and examining the structure under the microscope, it was noted that the tooth tips etched much more rapidly than the body of the tool, a fair indication of softening of the tips in grinding. Confirmation was obtained by carrying out a

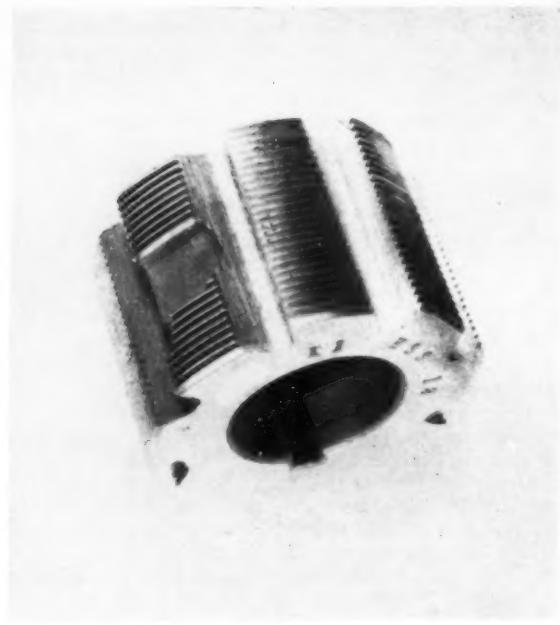


Fig. 3.—Similar hob in which the cracked portion lifted out.

series of hardness tests on the tool body and teeth, using a $2\frac{1}{2}$ kg. load on the Vickers hardness tester. The hardness values obtained were: dark etching tooth tip—685; lighter etching zone—890.

The foregoing examination covered a tool which had received the correct heat-treatment prior to grinding the teeth, and the defects arising during the grinding were not influenced by poor treatment, but can be considered as due to incorrect grinding technique.

Incorrect Heat Treatment

Several instances have also been encountered where the grinding hazards have been increased by the omission of tempering (secondary hardening in the case of high

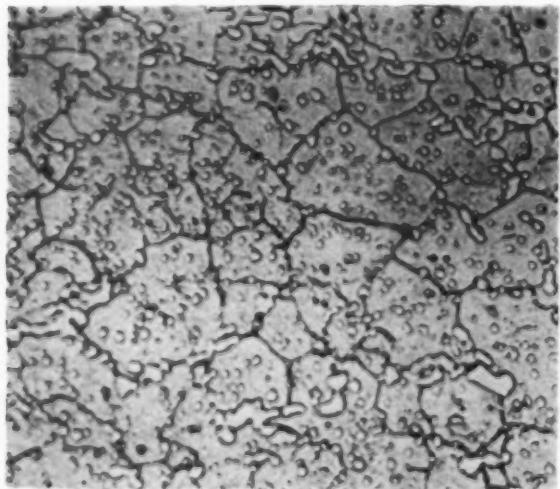


Fig. 4.—Structure of high speed steel hob which cracked during grinding. $\times 1000$

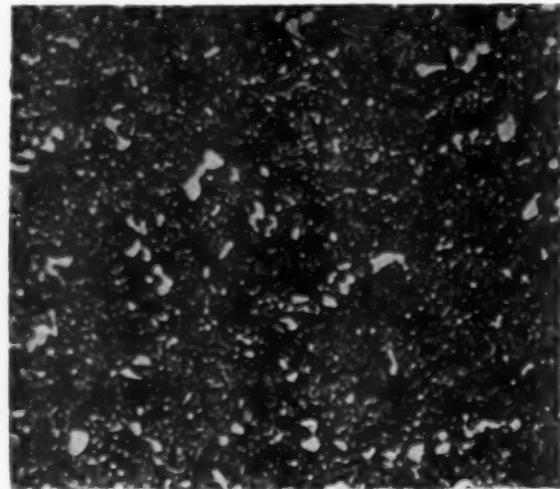


Fig. 5.—Structure of the same tool after tempering at 580°C. $\times 1000$

speed steel. In such cases, the heat developed during grinding will effectively secondary-harden or temper the heated portion, the volume change normally incident upon tempering being added to that of heat expansion, and grinding cracks will almost certainly form. Fig. 4 shows the structure of a hob which cracked during grinding and Fig. 5 concerns the same tool after tempering at 580° C., the hardness increasing from 63-65 Rockwell "C" as the result of the tempering.

Whilst it is possible to grind a tool in the hardened and untempered state with a Rockwell hardness in the normal range, the conditions will be more exacting than with a correctly treated tool. Conditions laid down and used for any particular grinding job will often be quite unsuitable if a previous operation, such as tempering, has been omitted. Hardness tests are, therefore, no real guide to the effectiveness or correctness of heat-treatment.

In the case of high speed steel, the necessity for tempering before grinding is particularly important, and the same applies to other alloy steels where a certain amount of austenite is retained after quenching. The breakdown of austenite to martensite, which occurs during tempering (secondary hardening), is accompanied by an expansion, and it is obviously desirable that this expansion should proceed as evenly as possible throughout the tool, any sudden changes in temperature being undesirable. Fig. 6 shows the usual form of the dimensional changes which take place on tempering high speed steel, and it will be obvious that any work on the tool involving the generation of heat (such as grinding) should be done after the tool has been allowed to go through these changes (by tempering), and not before.

Grinding an untempered high speed steel tool will thus tend to cause expansion of the surface due to the metallurgical change as well as the normal physical expansion effects of heat, and the risk of grinding cracks is greatly increased.

Perhaps the most difficult condition, as far as grinding is concerned, is reached when a tool has been overheated in hardening, for two main reasons:

- (1) A large amount of austenite is retained, and if the tool is not adequately tempered it is almost certain

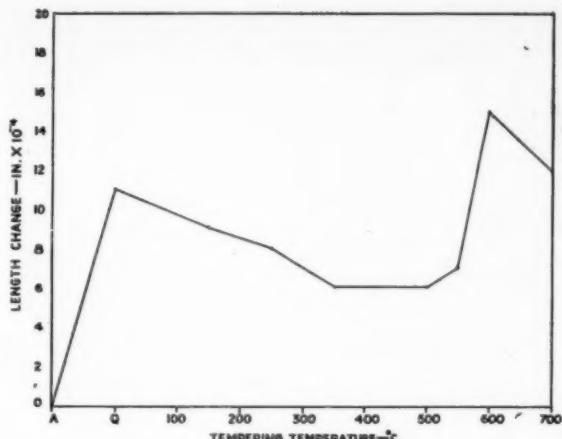


Fig. 6.—Change in length on tempering high speed steel.

that cracking will develop on grinding, due to the severe expansion effects.

- (2) The structure of the steel is coarse grained and it is more brittle than with correct treatment. The reduced toughness produces an added tendency towards cracking, and the tool may, in fact, have microscopic cracks already present within the grains of martensite.

Fig. 7 illustrates the structure of a high speed steel tool which had been overheated in hardening, and which cracked during grinding. The well developed grain boundaries form easy paths for cracking, and this type of structure is very undesirable. Fig. 8 shows the same tool after annealing and correctly heat treating.

In all the cases so far described, the effect of poor grinding technique, possibly combined with an unsuitable metallurgical condition, is to produce localised softening of the cutting edges of the tool. It is probable that the softened layer reaches a temperature in the region of 650° C. during grinding, but, as there is almost bound to be a temperature gradient from the immediate surface to the body of the tool, the extreme surface may reach a

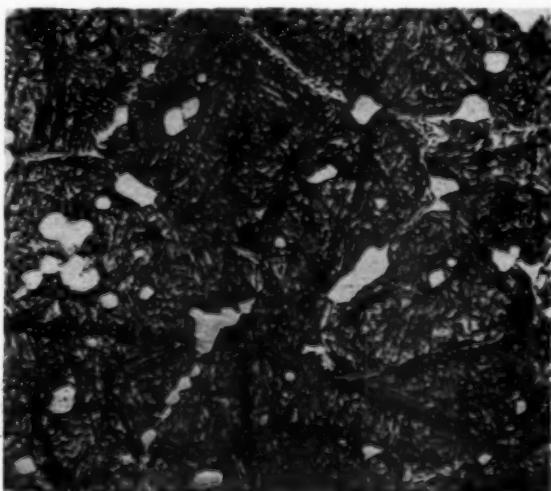


Fig. 7.—Structure of a high speed steel tool, overheated in hardening, which cracked on grinding. $\times 1000$

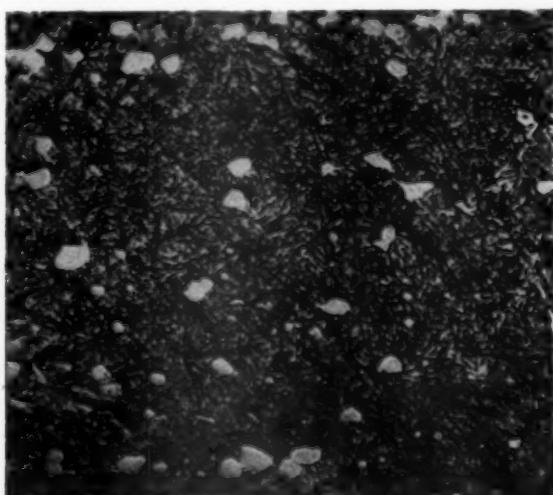


Fig. 8.—Structure of the same tool after annealing and heat treating correctly. $\times 1000$

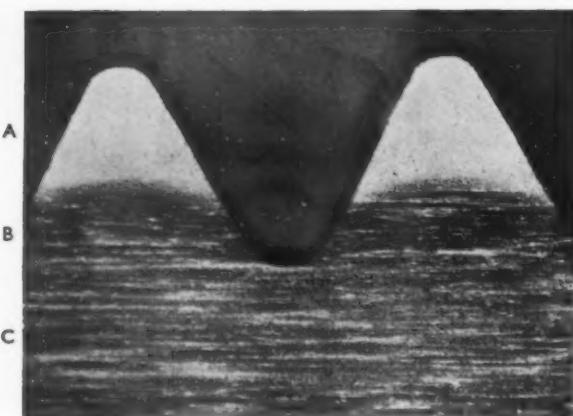


Fig. 9.—Etched section through hob teeth which had been re-hardened at the tips as a result of the temperature reached in grinding.

much higher temperature. In most cases, the extreme outer layer will be removed during the final grinding operation and a normal inspection will show a satisfactory surface with no "temper" colours or other indications of local heating. The untouched backs of the teeth, however, will usually show the temper colours quite strongly if the tool was not correctly ground.

Local Hardening due to Grinding

It has been assumed that the heat developed during grinding is, at the most, equivalent to a tempering temperature of about 650°C. (ignoring the extreme surface), but there is evidence to prove that, in an extreme case, the localised heating can be so intense as to reharden the tool surface. Such a condition existed in a hob which had several teeth cracked and lifted out following grinding, but in this case we failed to note any local softening of the teeth by the file test mentioned previously. It was observed, however, that the untouched backs of the teeth (forming the surface of the longitudinal gashes) were faintly marked in a manner similar to that following hardening, this pattern not being present on other unground parts of the tool.

One of the sections of teeth which had lifted out was examined under the microscope, and this showed the reverse of the condition normally observed when heavy grinding has occurred. The tooth tips resisted the etching reagent, whereas usually they etch more easily than the body of the teeth, and this resistance to etching is a good indication of the hardened and untempered condition. A dark etching zone was noted near to the roots of the teeth, the general pattern being as shown in Fig. 9.

It was considered possible that zone A (Fig. 9) represented an area of the tooth which had been heated sufficiently high during grinding to reharden it; that zone B was heated to a lower temperature within the softening range; and that zone C was unaffected. For this theory to be correct, then zone A should soften on tempering and the other zones remain unaltered up to the original tempering temperature.

Progressive tempering was applied to a portion of the teeth and a careful hardness test made after each temperature in all 3 zones: the results obtained are set out in Table I. It will be seen that the hardness in zone A falls quite sharply on tempering, an indication that

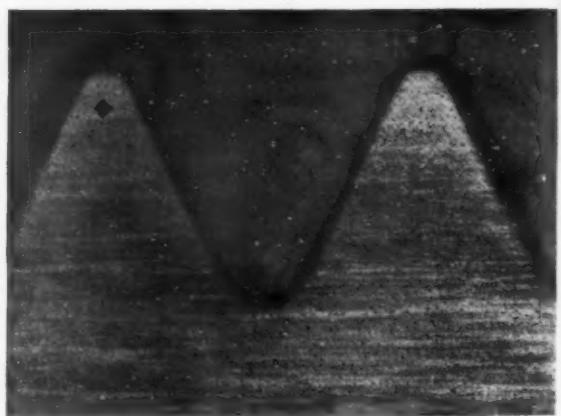


Fig. 10.—Etched section through teeth from same hob which had been re-tempered at 600°C.

this part was in the partially hardened condition. Zones B and C remained essentially unchanged in hardness, as would be expected.

TABLE I.—RESULTS OF PROGRESSIVE TEMPERING OF HOB TOOTH

Conditions	Hardness (V.P.N.)		
	Zone A	Zone B	Zone C
As received...	857	766	841
Tempered 330°C...	857	752	857
Tempered 400°C...	795	752	857
Tempered 550°C...	795	752	857
Tempered 600°C...	781	—	825

The microstructure of a tooth which had been re-tempered from 600°C. showed a uniformly etched surface, the original light etching tooth tip now responding normally to the acid etch. The structure is shown in Fig. 10.

There would appear to be little doubt, therefore, that in the case of this latter tool the grinding operation had been very badly performed, sufficient heat being developed to reharden the tooth tips. It is hardly surprising that cracking developed on the teeth, followed by lifting of the surface.

Conclusion

In this short article, an attempt has been made to indicate some of the factors responsible for the formation of grinding cracks, and, it will have been noted that in common with other manufacturing processes, there are several variables which can affect the response of the steel surface during the grinding operation. Many tools fail from grinding cracks without the user being aware that cracks existed, and the keen attention which is paid by some tool and component makers to ensure that no cracks are present indicates the danger existing.

Cracking usually develops when an unsuitable wheel or speed is used, or when an undesirable metallurgical condition is present. Some of the steels which require grinding are extremely hard and abrasive, necessitating the use of soft, open grained, wheels, and in all cases the wheel surface requires frequent dressing to remove lodged-in metal. Any tendency towards burning or discolouration of the work surface should be prevented by using:—

- (a) A softer wheel.
- (b) Decreased spindle speed, wheel diameter and width of wheel face.

(c) Increased work and traverse speed and feed.

It is also important that the work should not become stationary and in contact with the revolving wheel, otherwise glazing and cracking will probably take place.

From the metallurgical viewpoint there are several factors which can influence the quality of the finished work and the incidence of grinding cracks. In spite of the fact that most heat-treatment furnaces are equipped with pyrometrical control, or measurement, of temperature, many tools are overheated in hardening, and in many cases the temperature exceeds the correct figure by 100° C. from a nominal 800° C. The resulting structure is coarse grained and inherently brittle, so that there is little chance of successful grinding being done,

even under good conditions, and cracking almost invariably takes place.

Omission of tempering is also not uncommon, and many toolmakers are apparently still under the impression that high speed steel does not require any treatment other than hardening. If tempering (or more correctly secondary hardening), is omitted, then the ground surface will, as a result of the heat developed in grinding, go through a metallurgical change involving expansion, and cracking is likely to take place.

Acknowledgments

The author is indebted to the directors of Messrs. Arthur Balfour & Co., Ltd., Sheffield (in particular The Hon. F. H. Balfour), for permission to publish this paper.

Personal News

MR. E. F. FARRINGTON, at present Senior Civil Engineer, Maintenance, has been appointed Assistant to the Chief Engineer of the Appleby-Frodingham Steel Co., for duties in connection with capital expenditure schemes and developments; MR. R. B. ATKIN, Assistant Civil Engineer, has succeeded Mr. Farrington as Senior Civil Engineer. On April 1st, MR. K. PATERSON, Director and Chief Engineer of the Company, will take up the appointment of General Manager of The United Steel Structural Co., of which MR. H. SAUNDERS will be Commercial Manager and Director. MR. L. GASKELL, Chief Electrical Engineer, will become Chief Engineer to succeed Mr. Paterson, and MR. W. E. SMITH will become Chief Electrical Engineer and MR. D. R. M. NISBET, Deputy Chief Electrical Engineer.

MR. H. P. FORDER, Assistant General Manager (Commercial and Administration) of Samuel Fox & Co., Ltd., has been appointed Deputy General Manager, and MR. R. J. BAVISTER, Assistant Commercial Manager, is now Commercial Manager. MR. J. H. GOODLAD will deputise for Mr. Bavister as required. MR. A. GRAYSON, Buyer, has been appointed Supplies Manager, and is succeeded by MR. J. MASON. MR. G. J. CUMMINGS is now Education Officer.

MR. V. SMITH has been appointed Development Engineer of the Workington Iron & Steel Co.

MR. P. TATLOW has been appointed Sales Manager, General Engineering Department, of Air Control Installations, Ltd.

MR. T. W. J. TEMPLE has been appointed Manager of the Middlesbrough Office of The British Thomson-Houston Co., Ltd., in succession to MR. G. LLOYD JONES who is shortly taking up another appointment in the B.T.H. organisation.

MR. W. T. KITCHING is now associated with Campbell, Gifford & Morton, Ltd., who are constructing the first continuous casting plant at a Sheffield steelworks.

MR. T. W. WHITING relinquished his position as Chief Metallurgist of the Screw Division of Guest, Keen & Nettlefolds (Midlands), Ltd., at the end of last year, to take up an appointment as Materials Consultant to the Royal Canadian Navy at Ottawa.

MR. V. PENDRED, Sales Director of Head, Wrightson & Co., Ltd., of Thornaby-on-Tees is now in India, where he is making an extended tour of the country with Mr. M. H. Jenks from the Calcutta Office of Balmer Lawrie & Co., Ltd., whom Head, Wrightson have just appointed as their Selling Agents in India.

THE British Steel Castings Research Association announce that MR. C. H. KAIN, Joint Managing Director of Lake & Elliot, Ltd., of Braintree, has been elected Chairman of Council. MR. KAIN succeeds MR. F. N. LLOYD, Chairman and Managing Director of F. H. Lloyd & Co., Ltd.

MR. G. H. LATHAM, Chairman and Managing Director of the Whitehead Iron and Steel Co., Ltd., has been appointed President of the British Iron and Steel Research Association in succession to Sir Ellis Hunter, Chairman and Managing Director of Dorman, Long & Co., Ltd., who has occupied the office since 1945. MR. A. G. STEWART has been appointed President-Elect.

It is announced by The Plessey Company, Ltd., that MR. J. A. CLARK and MR. M. W. CLARK have been appointed Directors of the Company. MR. J. A. SMITH, formerly a Director, has retired from this office owing to ill health. It is also announced that MR. W. F. CROSS has been appointed Joint Secretary of the Company with MR. E. H. OUSTON.

MR. G. W. P. PAGE has retired from the staff of the Publicity Department of The British Thomson-Houston Co., Ltd. Mr. Page joined B.T.H. in 1923, having previously held Editorial appointments on *Electrical Engineering* and in the G.E.C. Publicity Dept.

NEWMAN INDUSTRIES, LTD., announce the appointment to the Board of MR. C. BUSH, Company Secretary, and MR. G. REDFERN, Chief Engineer.

MR. L. W. H. REA, who was recently appointed General Manager of Edwin Danks & Company (Oldbury), Ltd., has now been elected to the Board of Directors, following the resignation of CAPT. (E.) GREGSON, R.N.R. MR. W. SMITH, Works Manager, has also been elected a Director, in place of MR. J. W. HARGREAVES, who has also resigned.

SIR HARRY BROWN, C.M.G., M.B.E., who in 1940 joined the British General Electric Co., Pty., Ltd., of Australia as Chairman and Joint Managing Director, and later held the post of full-time Chairman, has resigned at the age of 75. MR. T. E. MORGAN, at present Managing Director, has been appointed Chairman and Managing Director of the Australian Company. At the same time, MR. A. T. BRIDGE, MR. R. J. DARTON and MR. C. DRABBLE have been appointed Directors.

MR. E. S. WADDINGTON of Philips Electrical Ltd., Industrial Products Department, has been appointed Hon. Treasurer of the Society of Engineers, which this year celebrates its centenary.

MR. F. HOWITT has relinquished the position of Technical Superintendent and Deputy Manager of the Birmingham Foundry and Forge of Northern Aluminium Co., Ltd., on being transferred to a sister company, Aluminium Co. of Canada, Ltd., Montreal. MR. S. WILDING has been appointed Production Manager, being succeeded as Works Engineer by MR. D. A. ELLIS. MR. A. R. MARTIN assumes the full responsibilities of Chief Metallurgist.

It is announced that MR. W. H. TAIT, Technical Director of The Glacier Metal Co., Ltd., for private and personal reasons has requested the Board to agree to his retirement from the Board and that the Directors have agreed to this. Mr. Tait will continue serving the Company in the technical field.

THE Directors of Thos. Firth & John Brown, Ltd., have appointed MR. A. BROOMHEAD to the Board. Mr. Broomhead was appointed a Local Director in January, 1952, with particular responsibility for the Departments manufacturing Rolled Products.

TUBE INVESTMENTS, LTD., announces the appointment of MAJOR-GENERAL R. F. BELCHEM, C.B., C.B.E., D.S.O., for special duties as a member of the personal staff of the Chairman, Mr. Ivan Stedeford.

SIR ALEXANDER DUNBAR has resigned from the Board of High Speed Steel Alloys, Ltd., but his services continue to be available to the Company in connection with their mining interests overseas. MR. W. D. PUGH has been elected a Director of the Company.

MR. W. J. MILLAR, who was Managing Director of Efco Engineering Co., Ltd., Joint Managing Director of Electric Resistance Furnace Co., Ltd., and a Director of Electric Furnace Co., Ltd., and Electro-Chemical Engineering Co., Ltd., retired on the 1st January, having reached retiring age.

DR. A. H. LECKIE, formerly Head of the Steelmaking Division of the British Iron and Steel Research Association, has taken up the position of Technical Officer with the Iron and Steel Board. DR. J. PEARSON is temporarily taking over the duties of Head of the Steelmaking Division, in addition to his present duties as Head of the Chemistry Department.

MR. D. LUTHER PHILLIPS retires under age limit from his position as Head of B.I.S.R.A.'s South Wales Laboratories on February 10th. He is to be succeeded by MR. S. S. CARLISLE, at present Head of the Instruments Section of B.I.S.R.A.'s Physics Department. B.I.S.R.A. will continue to have the benefit in a consultative capacity of Mr. Phillips's wide and intimate knowledge of the South Wales steel industry.

At the Company's staff dance, MR. AMBROSE FIRTH, Chairman of The Brightside Foundry and Engineering Co., Ltd., was presented with a long-service certificate marking his 40 years with the firm.

MR. D. I. KING, Assistant Editor of *The British Steel-maker* since 1950, has been appointed Editor. MR. G. R. LEWINGTON remains Managing Editor.

On the recommendation of the Managing Director, the Board of Newton Chambers & Co., Ltd., have appointed three Assistant Managing Directors, each with special responsibilities. They are: MR. P. J. C. BOVILL (Chemicals Branch), MR. S. C. TYRRELL (Excavator Branch) and MR. K. E. WALKER (Engineering Branch, including the Heavy Constructional Department and the Light Castings Department). The Company is also

strengthening its London organisation by giving Mr. R. F. A. Sampson (a Local Director) special responsibility for London interests.

MR. H. WILLIAMS, M.B.E., T.D., has been appointed Manager of the Worcester Branch of British Insulated Callender's Cables, Ltd., in succession to the late Mr. C. H. Panting.

Obituary

WE regret to record that Captain A. C. Jessup, Chief Metallurgist of Magnesium Elektron, Ltd., died in Rome on December 8th.

Captain Jessup, who throughout his professional life had been associated with the magnesium industry in England, France and Italy, and who had himself devised a process for the electrolytic production of magnesium, joined the Company in 1940 and became its Chief Metallurgist in 1947. He directed with great ability and foresight the research team which first brought to commercial development the magnesium-zirconium alloys, including the newer alloys containing thorium. He was an avid reader of the technical literature and probably the best informed metallurgist on magnesium.

Captain Jessup was a man of wide attainments both as a scientist and humanist. He combined great scientific gifts with a catholic taste for music, in which he took a profound delight. He wrote poetry. At heart a philosopher, he latterly devoted great effort to the study of nuclear physics and cosmology and made predictions in these fields subsequently verified by experiment. But to these intellectual attributes he also added the qualities of an arresting and original personality, and, by his gracious charm, warm humanity, and never failing sense of humour he endeared himself to a very wide circle of friends and colleagues at home and abroad.

WE regret to record the sudden death of MR. W. E. Benbow, Editor of *Iron and Steel* since 1935. He was taken suddenly ill on the evening of December 28th, and died the next day without regaining consciousness.

Born in Kent in 1900, Mr. Benbow began his technical career in 1917 in the metallurgical laboratories of Vickers, Ltd., Crayford, leaving them to be assistant to the chief draughtsman in the aircraft production section of Hooper & Co., Ltd. From there he went to J. & E. Hall, Ltd., Dartford, where he was engaged on work on foundry control, the routine control of engineering materials and the investigation of metallurgical problems associated with refrigeration. He joined the staff of *The Tramway World* in 1928 and a year later entered the service of Louis Cassier Co., Ltd., as assistant editor of *Iron and Steel Industry*, as it then was, and *Metal Industry*. In 1933 he became the first editor of *The Welding Industry*, now *Welding and Metal Fabrication*.

Mr. Benbow was a man of many parts. He spoke both French and German and had a good working knowledge of Italian, Spanish and Portuguese. His store of general knowledge was amazing, and when at short notice he was called upon to provide the important introductory chapter to the book "Steels in Modern Industry," he contributed a section which received high praise from expert critics; all this in addition to his work as general editor of the contributions of the other twenty-eight specialist authors.

Isothermal Heat Treatment

With Special Reference to American Practice

Mr. L. Rousseau, Vice-President of the Ajax Electric Corporation of Philadelphia, recently lectured on American isothermal heat treatment practice in London and a number of provincial centres. Before certain of these lectures, Mr. J. McMullen, Chief Metallurgist of Electric Resistance Furnace Co., Ltd.—an associate of the Ajax Company—gave an introductory talk on the principles behind the techniques discussed. An abridged version of the combined lectures is presented here.

IN 1929, Lewis, working in England, showed that a 0.80% carbon steel quenched into molten salt at 230° C. was still in the austenitic condition and could be bent quite easily. Further, on subsequent cooling, martensitic hardening occurred whether the cooling was rapid or slow. A continuous quench from austenitising (hardening) temperature to room temperature was not, therefore, essential for full hardening.

In 1930, Davenport and Bain published their classic work in the U.S.A. showing that the transformation of austenite could be effected at constant temperature providing sufficient time at temperature was allowed. The curves showing the commencement and completion of transformation were referred to as "S curves." They are now more commonly called Time-Temperature-Transformation diagrams or simply "TTT curves." Curves for most of the heat treatable British steels have now been published.

Fig. 1 represents a TTT diagram for a 1% carbon steel, with time plotted on a logarithmic basis. The more important features of the diagram are as follows:—

(1) At temperatures just under the lower critical, there is a long delay before the start of transformation. This "time delay" or "induction period" decreases with transformation temperature, reaching a minimum at the "pearlite nose" or "knee"—in this case about 550° C.

(2) At temperatures just under the critical, the transformation product consists of large globules of carbide in ferrite, the structure becoming more lamellar and finer, and, therefore, harder as the transformation temperature is lowered.

(3) Below the nose, the time delay again increases. The products in this range, which are of intermediate hardness, are referred to as "intermediate structures" or bainite. As before, hardness increases with decreasing temperature of transformation.

(4) At the temperature marked M_s , in this case at 180° C., a transformation commences which is no longer isothermal. This is the formation of martensite which, once started, continues in proportion to the fall in temperature. The temperature at which transformation can be completed is marked M_f on the diagram.

The shape of the TTT curves depends primarily upon the composition of the steel, but in any given case is also governed by other factors, like austenitising temperature, austenitising time and grain size. The general effect of carbon and most alloying elements is to move the curve to the right—thereby increasing the induction period—and to lower the M_s temperature. In some cases a second minimum may appear below

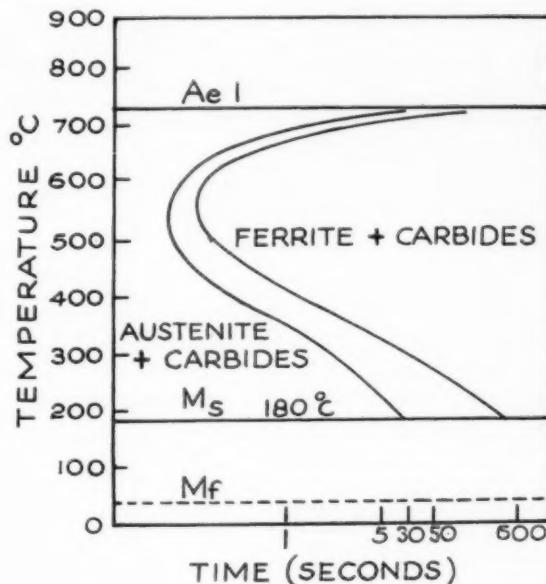


Fig. 1.—Time-temperature-transformation diagram of a 1% carbon steel.

the pearlite nose—the so called bainite chin—and this may or may not protrude beyond the nose.

With the relevant TTT curve as a guide, it is often possible to replace the conventional heat treatment or annealing cycle by an isothermal treatment, the three main types, which will be considered in greater detail later, being isothermal annealing, austempering and martempering. In each of these, the steel is first austenitised as in conventional heat treatment, by heating to the appropriate temperature, prior to quenching in a salt bath which is held at the appropriate transformation temperature. Salt baths for use in such operations have a number of interesting constructional features.

Quenching Baths

Not only does the furnace need a heating system to melt the salt and to bring it to the desired operating temperature, it also requires a cooling system and both of these must be automatically controlled. For this type of work, the furnace almost always has a metal pot, the outside of which is frequently finned. To extract the heat from the furnace, air under pressure from a motor driven blower is forced around the outside of the pot. Furthermore, in quenching work, the salt must be

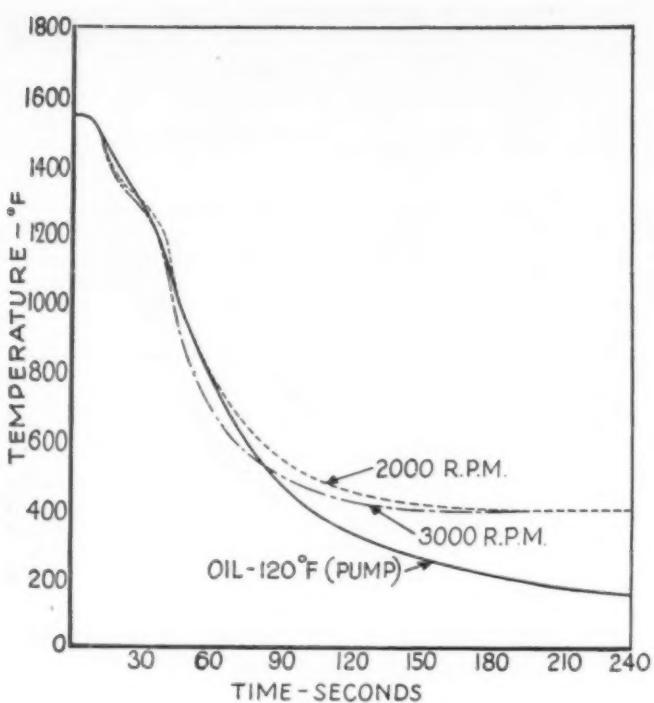
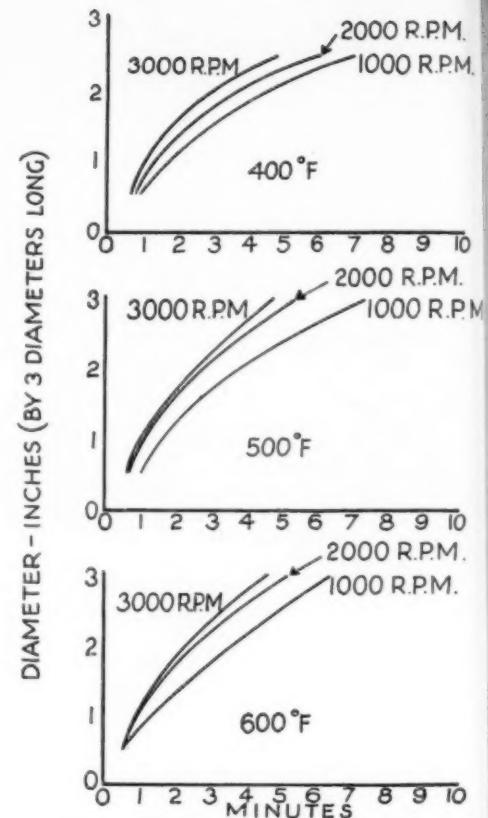


Fig. 2 (above).—Comparative quench curves for oil and salt.

Fig. 3 (right).—Effect of quenching temperature and salt agitation.

properly agitated, and salt pumps, with variable output, are provided for this purpose. The effect of the motion of the salt in extracting heat from the work is not always appreciated, but it is very pronounced, and, with proper design, will actually permit a salt quenching furnace operating at elevated temperatures to extract heat from the work, over the critical range, at a rate equal to or greater than that which can be secured from an oil quench tank equipped with equivalent circulating pumps. This is illustrated graphically in Fig. 2 from which it will be noted that at maximum agitation the rate of cooling is greater than that of oil. Fig. 3 illustrates the effect of quenching temperatures and of agitation on pieces of various diameter. The circulation is measured arbitrarily by pump speed, and it will be seen that the quenching power is little affected by a variation of 110° C. of the quench bath temperature, although, of course, the resulting hardness will be.

A last major requirement exists in the quench furnace. If the work is heated in salt, as is generally the case, a certain amount of high temperature salt will be carried over from the heating furnace into the quench furnace. This salt, generally a mixture of sodium and potassium chlorides is sparingly soluble in the quench salt, which is normally a mixture of potassium nitrate, sodium nitrate and sodium nitrite. The higher the temperature, the higher the amount of chlorides held in solution in the nitrates, and an excess of chlorides over the amount held in solution drastically decreases the quenching power of the quench salt. Therefore, a well designed furnace must have an effective automatic salt purifying system. This is secured by building the furnace with two separate chambers. One is the quenching chamber,



in which the pumps are located, and the other is the separating chamber, which is generally operated at a temperature some 30–50° C. lower. The chlorides are thus precipitated out of solution, and, after filtration, the cleaned salts are returned to the quenching chamber. Fig. 4 illustrates a modern quenching furnace embodying all the features just described.

Isothermal Annealing

In this technique, transformation is effected in the pearlite zone—usually in the range 620–700° C.—for the purpose of softening prior to machining or cold forming. It is particularly applicable to those steels where the time for transformation is short enough to offer some advantage over full annealing by furnace cooling.

The temperature selected for austenitising markedly influences both the transformation time and the type of pearlite formed. Consequently, it is often possible to shorten the transformation time by austenitising at a temperature lower than that normally selected for hardening. In the same way, a spheroidised structure, often preferred for optimum machinability in the higher carbon steels, is favoured by a low austenitising temperature. The time-temperature relationships for isothermal annealing are shown in Fig. 5.

As the transformation temperature is above the pearlite nose, a quick quench is not essential, although care must be taken to see that cooling is not so slow as to allow any part of the load to fall below the selected transformation temperature. The absence of critical requirements in quenching has led to some recent developments in the United States whereby "cyclic annealing" is carried out straight from the forging

hammer without re-austenitising, thus utilising the residual forging heat. In practice, the forgings are transferred direct from the hammer to the salt baths, held for the time necessary to complete transformation, and then cooled in almost any way, water quenches and sprays having the advantage of removing scale present before transformation.

From a practical standpoint, this operation has certain limitations :—

(1) The chemistry of the steel involved should permit relatively short annealing cycles, generally one hour or less.

(2) In order to be economical, the technique is usually combined with a forging operation.

(3) It is practically limited to a wholly new installation, because the annealing furnace must be installed in the forge shop, adjacent to the hammer or press.

(4) A production run of the continuous type for a relatively long period of time is required, and the characteristics of the steel cannot be changed except in a very minor way.

On the other hand, its advantages are as follows :—

(1) A very short annealing cycle is possible—less than one hour.

(2) A clean surface is generally obtained, because no air cooling after forging occurs, and because a water quench is used after the anneal.

(3) A uniform structure having superior machining properties is produced.

(4) Under proper conditions, the cost of operation is very low.

Because of the limitations previously outlined, cyclic annealing installations are not common. Nevertheless, those currently in service are outstandingly successful. The biggest installation is at the Ford Motor Company's Forge Plant in Canton, Ohio, where four mechanised units are used, two for annealing ring gears and two for pinion gears. The ring gear installation anneals induction-heated forgings of SAE-4028 (C, 0.25-0.30%; Mn, 0.70-0.90%; Si, 0.20-0.35%; Mo, 0.20-0.30%). The forgings are transferred directly from the press to the furnace, the temperature of which is 645°C., and the time cycle is 45 minutes. After removal from the furnace, the forgings are flash quenched in hot water for a few seconds to strip off the salt and all the slight scale present on the piece: no sand or shot blasting is used.

In the pinion gear units the set-up is the same, but the steel is SAE-8620 (C, 0.18-0.23%; Mn, 0.70-0.90%; Si, 0.20-0.35%; Ni, 0.40-0.70%; Cr, 0.40-0.60%; Mo, 0.15-0.25%). The temperature is again 645°C. and the time is 45 minutes.

Several other parts originally cyclic annealed have reverted to conventional practices because, in practice, the economies expected were not found to exist. This was largely because a reheat was involved, due to the shape and weight of the pieces, and because of the high standby losses of the annealing furnace when not in use.

Another installation which has operated successfully for several years is at one of the U.S. Steel Company's plants for the manufacture of the motor body of the 3.5 in. bazooka. This body is produced from tubing of SAE-4140 (C, 0.38-0.43%; Mn, 0.75-1.0%; Si,



Fig. 4—Modern isothermal quench bath with centrifugal pump and salt-separating chamber.

0.20-0.35%; Cr, 0.80-1.10%; Mo, 0.15-0.25%) approximately 2½ in. diameter and ½ in. wall, and is forged at one end to provide the Venturi for the propulsive gases. The tube is selectively heated in salt to a temperature of 1,175-1,200°C., the annealing temperature being 650°C. and the time cycle 30 minutes. The work is handled in baskets, the baskets being in the furnace when loaded so that no serious temperature drop is involved. Hardness as annealed is Rc 17-19*. Despite the fact that only a portion of the piece is heated and annealed, very uniform machining properties are secured in the entire piece.

A third installation of considerable magnitude is in operation for annealing after forging of 105 mm. artillery shells. The steel in this case is SAE-1012 (C, 0.10-0.15%; Mn, 0.30-0.60%). The temperature is 620°C. and the time cycle is 10 minutes. No operating experience has yet been made available on this installation.

In general, it may be said that cyclic annealing has very limited application: where all factors are favourable, it is high practical, but for a general purpose forging plant, it is not applicable.

* Rc indicates Rockwell "C" Scale.

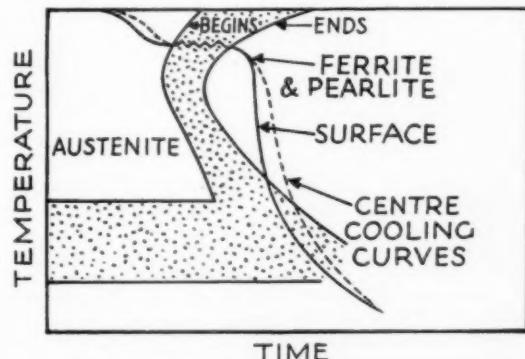


Fig. 5.—Diagram illustrating time-temperature relations for isothermal annealing.

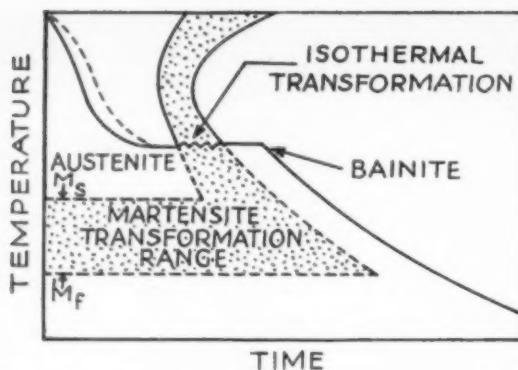


Fig. 6.—Diagram illustrating time-temperature relations for austempering.

Austempering

This is perhaps the earliest of the isothermal techniques, and consists of producing a structure wholly or substantially of bainite when intermediate hardness with a high degree of toughness is required. The steel is first austenitised as in conventional hardening. It is then quenched into a bath of molten salt held at the selected temperature—normally between 250 and 350°C.—and, after allowing sufficient time for transformation to be completed, withdrawn and cooled to room temperature. Fig. 6 depicts the procedure.

It is important that, despite the relatively high temperature of the quench, the steel is cooled sufficiently rapidly to avoid transformation at the pearlite nose. It is also important that the holding time is long enough to allow transformation to be completed at the selected temperature. The procedure is, therefore, limited to relatively thin sections of carbon steels, or to those alloy steels which, while possessing sufficient hardenability to preclude high temperature transformation, do not require a long transformation time. In this respect, it should be noted that, because of variations in transformation characteristics known to exist in commercial steels, it is necessary to allow a holding time about double that shown by the relevant TTT curve.

The benefits to be obtained by using the austempering technique are:—

(1) A bainite structure is produced without involving the martensite change and its attendant stresses: distortion is therefore, reduced or eliminated.

(2) The mechanical properties are superior to those obtained by the conventional quench and temper—at least in the case of the higher carbon steels.

A modification of the conventional austempering treatment has become largely accepted in the United States, and has greatly expanded the field of application. It is called the three-step austempering treatment, as compared to the conventional, which is two-step. In the three-step operation, the quenching temperature is not determined by the final hardness desired. Instead, it is lowered appreciably to permit much heavier sections to be quenched or, for the same section, to permit the use of

lower hardenability material. Hard bainite is produced which is subsequently tempered to the desired hardness. With this type of treatment, better physical properties for most purposes are possible, while retaining just as good control of distortion as is available with the conventional austempering treatment.

Generally speaking, austempering offers the following advantages:—

- (1) Excellent control of distortion.
- (2) Greater toughness at equal hardness.
- (3) Better fatigue life.
- (4) Good appearance.
- (5) Lower manufacturing costs.

The technique is extensively used throughout American industry as will be exemplified by the following applications. Springs are, in general, an excellent application. In the majority of cases, they are produced by the two-step method. In the automotive industry, for instance, at the Delco Remy Division of the General Motors Corporation, four large installations are in use. The first was selected because of constant difficulties with a spring used in the signal light indicator (right and left turn) commonly used on most cars. A two furnace set-up, both salt baths, was installed several years ago, and completely eliminated previously encountered breakage. Subsequently, three large installations utilising atmosphere furnaces for heating and salt baths for quenching, have been placed in service for similar articles used in ignition and accessory devices. The steel is generally a carbon manganese type of the SAE-1000 series. The temperature used is between 270 and 370°C., and the time cycle varies with the work, which on occasion is carburised. As a specific example, the SAE-1024 cranking motor clutch, gas carburised to 0.025 in. depth, is quenched from 925°C. into salt at 265°C. and held for 2½ hours. A similar application can be seen at the United Carr Fastener Corporation in Cambridge, Massachusetts where a multitude of stampings requiring spring-like properties are produced.

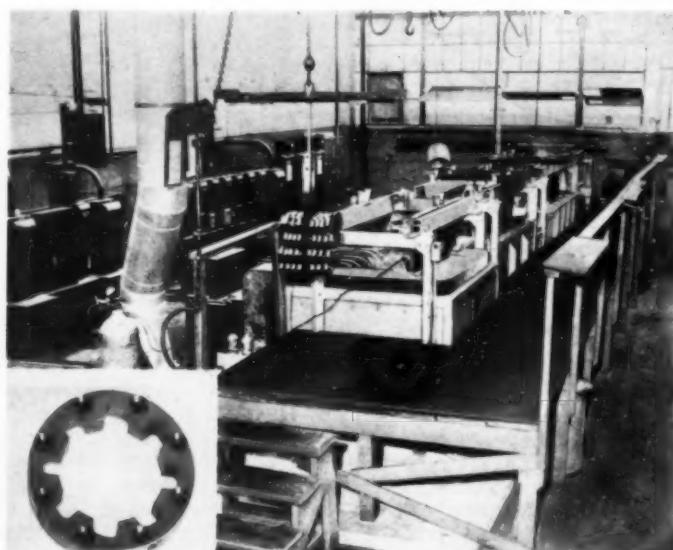
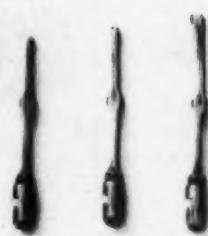


Fig. 8.—Battery of salt baths for austempering the Dynaflow cam (inset) by the three-step method.

LOCATION SERVICE FAILURES



Former heat treatment : no set at point of fracture.

Austempered : repeated shocks at point "A" would not cause fracture at corner "B"—point of failures.

Austempered : examples of set without fracture.

All components made from Steel W.D. 1075 and heat treated to the same hardness.

LOCATION SERVICE FAILURES



Former heat treatment : no fractures in endurance test.

Austempered : repeated shocks in two directions caused no fractures in spite of the extreme set.

All components made from Steel W.D. 1075 and heat treated to the same hardness.

BENDS WITHOUT FAILURES



Fig. 7.—Comparison of components heat treated to the same hardness by hardening and tempering and by austempering.

Manufacturers of business machines, such as International Business Machine, Remington Rand, etc., make considerable use of austempering. Of particular interest is the fact that the type bars can be austempered in a single shape and subsequently bent to the widely different requirements of the different letters. The high ductility of the austempered piece permits this very economical manufacturing technique. Some of the work is fabricated in SAE-C-1050 (C, 0.84-0.55%; Mn, 0.60-0.90%); and treated as follows: heat in salt to 830-840° C. for 4 minutes, quench at 330° C. for 16 minutes.

A widely accepted operation is that of austempering cast iron cylinder liners for gasoline and diesel heavy duty engines used in the trucking and allied industries. The three-step treatment is used in this case, and The Thompson Products Company of Cleveland, Ohio, the Sealed Power Company, Rochester, Ind., and the International Harvester Company of Chicago, Illinois, have large mechanised installations. The average composition of the iron is C, 2.95-3.25%; Si, 0.8-2.2%; Cr, 0.20-0.34%; Ni, 0.20-0.40%; Mo, 0.40-0.60%.

The treatment is substantially identical in the several plants, and consists in austenitising at 830-840° C. for approximately 12 minutes, quenching at 200° C. for 12 minutes, and tempering at 220° C. for 24 minutes. The resulting hardness is in excess of Re 50. The main object, however, is to keep the distortion within such low limits that all finish machining and grinding can be done before heat treatment, with only the final honing operation taking place after hardening. The distortion incurred is consistently kept below 0.005 in. out of round, on an average.

Brake shoes and sway bars, such as are used in the

current models of the Ford Motor Company, are being austempered. The sway bar is an especially interesting application, in that the bar is heated for forming to a temperature of the order of 1,100° C., hot formed, and mechanically transferred to an austempering furnace equipped with an internal chain type conveyor. This installation is still having some slight mechanical difficulties, but produces a very satisfactory product very economically. The steel is SAE-1085, the quench temperature 270° C. and the time of transformation 10 minutes.

Many other applications exist which are too numerous to cover in detail here. They include shoe shanks (the original application), clicker (shoe stamping) dies, power saw guides, and screw machine parts, whilst wire patenting is a version of the technique.

Martempering

This process consists of transferring the austenitised steel to a bath held at a temperature just above the M_s temperature, holding long enough for temperature equalisation, then cooling in air. Normal tempering follows.

With the temperature above the M_s temperature for the steel concerned, the thermal gradient and resulting stresses are dissipated while the steel is in the soft austenitic condition and as the transformation to martensite occurs during the subsequent air cool, transformation stresses will be at a minimum and distortion reduced or eliminated. A certain amount of self-tempering may occur during the air cool.

As in austempering, the conditions of the quench must ensure a cooling rate fast enough to avoid transformation

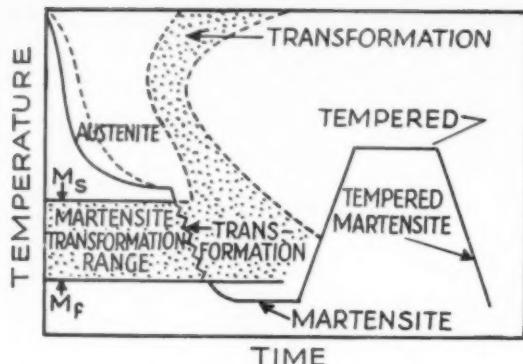


Fig. 9.—Diagram illustrating time-temperature relations for martempering.

at the higher temperatures. A further limit is that the time required for equalisation must be less than the induction period at the holding temperature selected, otherwise isothermal transformation to bainite will commence.

Although martempering theoretically calls for quenching at a temperature slightly above the M_s point, holding until stable temperature conditions are established and then air cooling, many practical applications have a quench temperature below the M_s point, in order to permit the handling of larger sections. In the tempered condition, the hardness with martempering is equal to that produced by the conventional quench and temper practice. As quenched and air cooled, but before tempering, it is lower than the oil quenched hardness because of partial tempering. The quench furnace requires effective and controllable agitation of the salt. Astonishingly large sections are currently being handled. For instance, 120 mm. armour piercing shot show Re 60 in the centre of the solid piece, and 6 in. thick castings used in mining machinery are martempered to a surface hardness of Re 60.

In martempering, the steel is usually much more highly alloyed and greater hardness is secured than in austempering.

Among current applications of interest are ball and roller bearing races made of SAE-52100 steels (C, 0.95-1.10%; Mn, 0.25-0.45%; Si, 0.20-0.35%; Cr, 1.3-1.6%). The main object of the treatment is to control the distortion created by hardening, and to provide a service performance at least equal, and often superior, to that of the conventional treatment. Several large bearing plants in the United States are using martempering in at least part of their production.

A wide variety of tools and dies are martempered in plants too numerous to mention, whilst other applications include: cast iron cylinder sleeves for the Detroit Diesel Division of the General Motors Corporation; cutlery, both of stainless and carbon steels; gears of alloy steels for mining machinery; handling devices; tractors, the main and auxiliary connecting rods of the Wright turbo-compound engine. The last-named is made of SAE-4350 steel (C, 0.48-0.53%; Mn, 0.60-0.80%; Si, 0.20-0.35%; Ni, 1.65-2.00%; Cr, 0.70-0.90%; Mo,

0.20-0.30%) and is tempered to a final hardness of Re 36-38.

One very interesting application has produced spectacular results from a performance as well as a low manufacture cost standpoint. It is the hardening of lawn mower blades, both of the reel type and of the rotary type. The steel is practically universally SAE-1065 (C, 0.60-0.70%; Mn, 0.60-0.90%), a cheap, relatively low hardenability steel. The treatment consists of austenitising at 860°C. for 5 minutes, quenching at 345°C. for 5 minutes, air cooling for 5 minutes and water quenching. The final structure is partly bainite, tempered martensite and untempered martensite. It is decidedly abnormal and, metallurgically speaking, probably unsound. However, its performance in use has been outstanding and more and more lawn mower manufacturers are adopting it.

Combined Carburising and Isothermal Treatment

The fact that some of the low-carbon low-alloy steels not normally amenable to austempering and martempering can be rendered suitable for such treatment by preliminary carburising opens a field of combined carburising and isothermal treatment. The control of distortion possible in this process is reported to have shown a reduction in overall time cycles in cases where straight carburising necessitated grinding after hardening.

The carburising may be done by any method selected, pack, gas or liquid, air cooled to room temperature, and subsequently reheated and martempered, or the work may be carburised and martempered directly from the carburising operation.

In the first case, martempering is being effected in the manner previously discussed. Examples of such applications are:—

(1) Gears of various types, especially highly stressed aircraft gears used in turbine-type engines, where both good core and case properties are required. In actual practice, approximately two-thirds of the gears involved in a popular model are martempered in this manner, with

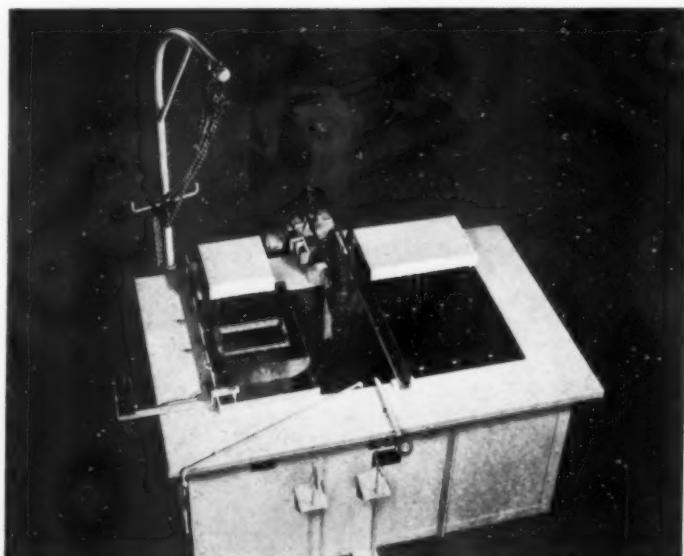


Fig. 10.—Martempering of bearing races eliminated the need for die quenching equipment.

distortion kept within specified tolerance. The balance are die quenched.

(2) Domestic refrigerator and air-conditioning unit parts, such as eccentrics and rollers for the Frigidaire Division of the General Motors Corporation.

The second technique, martempering directly from the carburising furnace, is, because of handling problems, largely confined to liquid carburising methods. Since it is not possible to quench directly from cyanide salts into nitrate salts, an additional furnace is used between the carburising and the martempering units. Its operating temperature is usually lower than the carburising furnace. Outstanding examples of parts treated in this way are:—

(1) Two-cycle motor crankshafts for the great majority of outboard engines used in the United States. The two largest producers, Evinrude & Johnson Motors are using this method throughout and other producers have adopted the technique. The steel used is SAE-4615 (C, 0.13-0.15%; Mn, 0.45-0.65%; Si, 0.20-0.35%; Ni, 1.65-2.00%; Mo, 0.20-0.30%).

(2) Heavy duty truck transmission gears made of SAE-4615, produced by the Wisconsin Axle Division of the Timken-Detroit Axle Company, are treated as follows: carburise, 925° C. for 4 hours; neutral wash, 1½ minutes at 815° C.; quench, 5 minutes at 340° C.; air cool-30 minutes and, finally, stress relieve at 180° C. for 30 minutes.

New and Revised British Standards

BRITISH STANDARDS FOR WORKSHOP PRACTICE (B.S. HANDBOOK NO. 2: 1953) PRICE 2s.

FOURTY-ONE of the most vital standards to engineers, to draughtsmen and to engineering students have been collected in the 1953 edition of this Handbook. Analogous standards have been grouped into seven sections, namely those dealing with limits and fits; screw threads; gauges; bolts, nuts and metal thread screws; cutting tools; machinery details; and miscellaneous data. In the foreword, Mr. S. J. Harley, Chairman of the Engineering Divisional Council of the Institution, expresses the hope that the Handbook will become increasingly recognized as the standard text book in all establishments where engineering is taught, both in industry and in the academic world. He also draws attention to the fact that the opportunity has been taken of including in this reprint several specifications, the basic data for which has received acceptance on both sides of the Atlantic, notably those standards for unified screw threads and for limits and fits.

80 AND 90 OZ. WINCHESTER BOTTLES (B.S.830: 1953) PRICE 2s.

In 1939 the B.S.I. issued a standard for Winchester bottles at the request of the Association of British Chemical Manufacturers and the British Chemical Plant Manufacturers Association. Because of the wartime difficulties of supply, the standard was not widely used, and a revision has now been prepared which is consistent with current practice. The new standard, B.S.830: 1953, 80 oz. and 90 oz. Winchester bottles, relates to plain and fluted containers, and in order to facilitate storage and outer packaging the same maximum diameter has been adopted for both types in each capacity. Alternative types of neck finishes are provided.

PACKAGING CODE : USE OF DESICCANTS IN PACKAGING (B.S.1133, SEC. 19: 1953). PRICE 3s.

The deterioration of the contents of packages is of concern to all packers, but in particular to exporters. Metal goods are liable to corrode in damp and humid climates, and materials such as textiles, leather and paper are subject to other forms of deterioration. Some articles can be given protective treatment during manufacture, and others can be protected by the application of corrosion preventives, and by other means. Some, however, such as delicate instruments, cannot be given

any such treatment, and for these it is essential to ensure that the humidity within the package remains reasonably constant.

The type of packaging which achieves this is known as a desiccated package, and this method is dealt with fully in Section 19, which describes the method and deals with factors governing the choice of desiccant, the determination of the quantity needed and other general packaging considerations.

The new section completes that part of the Packaging Code which deals with protection of contents against deterioration, the other relevant sections being: Section 5 (Protection against Spoilage by Micro-Organisms, Insects, Mites and Rodents. Price 2s.), and Section 6 (Temporary Prevention of Corrosion of Metal Surfaces during Transportation and Storage. Price 10s. 6d.).

METHODS FOR THE ANALYSIS OF ALUMINIUM AND ALUMINIUM ALLOYS (B.S. 1728. PART 5: COPPER (ABSORPTIOMETRIC METHOD)) PRICE 2s.

This British Standard Method for the absorptiometric determination of copper is intended for use with alloys containing smaller percentages of copper than those covered in B.S. 1728, Part 1 (of the order of 0.01-0.3%). The method specifies the reagents used, recommended methods of sampling and test procedure. An indication of the reproducibility expected is given and is derived from experiments carried out by a number of independent analysts.

RUST, ACID AND HEAT RESISTING STEEL WIRE FOR SPRINGS (B.S.2056: 1953). PRICE 2s. 6d.

THIS standard covers the same types of steel as B.S.1554, Rust, Acid and Heat Resisting Steel Wire, and specifies the requirements for such wire when it is to be used in the form of springs.

Copies of the Handbook and Standards, can be obtained from the British Standards Institution, 2, Park Street, London, W.1.

Denison Midlands Office

SAMUEL DENISON & SON, LTD., makers of weighing and testing machines, have opened a Midlands Office at the White House, 111, New Street, Birmingham, 2 (telephone: Midland 3931; telegrams: Weigh, Birmingham). Mr. B. A. Lewis has been appointed Midlands Manager.

Correspondence

CARBON IN THE ENGINEERING AND METALLURGICAL INDUSTRIES.

The Editor, METALLURGIA.

Sir,

I have read with interest the article "Carbon in the Engineering and Metallurgical Industries," by V. S. Kingswood, in the October, 1953, issue of METALLURGIA. In the introductory sentence to the section on industrial diamonds, he says: "Industrial diamonds are usually natural products, because synthetic stones are most difficult and costly to produce as specimens of workable size." This statement is not quite correct, however, as up to the present only natural diamonds have been available, and synthetic diamonds have never actually come into the trade. The author may rightly refer to the J. B. Hannay diamonds in the Natural History Section of the British Museum, but of these unique specimens the longest is only about 1 mm. in length. It is difficult to understand, therefore, where the author obtained the information that the synthetic diamond is most difficult and costly to produce as specimens of workable size. The literature, of course, abounds in claims of synthetic diamonds, but the diamond trade knows only the natural variety.

The statement that diamond mines are to be found in many parts of the world is also somewhat erroneous, as they exist only in the central and southern parts of Africa, and in the northern part of South America. Of course, diamonds have occasionally been reported in North America, Canada, Australia and India, but not in large quantities: for instance, the Indian mines are today almost exhausted. The author distinguishes between three varieties: (1) the transparent crystalline gem variety; (2) boart, an impure variety of (1); and (3) carbonado. Reference is then made to the fact that the second class is mainly used industrially. Here, the author follows the mineralogical classification of diamonds, which originally only recognised the gem variety and non-gem variety, i.e. boart. The trade definition, however, today considers as boart only the so-called crushing boart* used for producing diamond powder, which is used in bonded and loose form.

Yours faithfully,
P. Grodzinski.

London, 9th December, 1953.

The Editor, METALLURGIA.
Sir,

In reply to Mr. Grodzinski's letter, I should like to make the following general observations.

On reading the first paragraph, it is not clear to me what point is being made since in questioning my statement that "industrial diamonds are usually natural products because synthetic stones are most difficult and costly to produce as specimens of workable size," the writer goes on to state, in support of his argument, that "up to the present only natural diamonds have been available and synthetic diamonds have never actually come into the trade." This seems to me to imply that which I had already stated.

Later, on the same point, we find the writer refers to the unique specimens of Hannay in the British Museum, the largest of which is about 1 mm. in length, and also states that despite claims in the literature to the

manufacture of synthetic diamonds, the diamond trade knows only the natural variety, and yet disagreement is expressed with the point made above referring to the difficulty and cost of manufacture.

Surely the fact that synthetic specimens are not appearing on the market, is probably due to one of two main causes. Either they are uneconomic to produce in workable sizes for industrial requirements, or there is strong business influence amounting almost to a monopoly in the marketing of natural diamonds. Unfortunately, I know very little about the latter, whereas Mr. Grodzinski must have expert knowledge, but I feel sure that my first supposition is not very far from the truth.

The statement on the location of diamond mines was in every sense meant to be a generalisation and I have no hesitation in agreeing with the writer.

The trade classification submitted is interesting, but I was under the impression that "boart" may be subdivided into the more massive variety used for tool tips and drawing dies, and that which is finely divided, and used for grinding, polishing, etc.

In conclusion, I would like to thank Mr. Grodzinski for his interest in these articles and for information supplied.

Yours faithfully,
V. S. Kingswood,
Battersea Polytechnic,

London. 23rd December, 1953.

The Dechema Institute

LAST month, the Dechema Institut fur Apparate und Stoffkunde (Dechema Institute for Chemical Apparatus and Materials) was formally opened in Frankfurt-am-Main. Some 200 well-known personalities in scientific and industrial circles, including a number of foreign visitors, were present at this function, when Dr. Hans Broche, Essen, and Senator E. h. Dr. Bretschneider, Chairman and Managing Director of the Dechema respectively, made some illuminating remarks concerning the aims and objects of the Institute. The Institute represents an attempt to illustrate in a practical manner the broad area covered by the field of chemical apparatus and equipment. Prof. Dr. Grassmann of Zurich reported on the success that he had with a series of lectures, which ran through several semesters, on the subject of chemical apparatus, equipment and materials. The Dechema Institute will classify in one register (the Dechema Register) all chemical apparatus and equipment with respect to its theoretical bases of design, its construction, mode of operation and application to industry. A compilation of 100 materials with full particulars of their physical properties and resistance to 1,000 active chemical agents is now appearing in a Third Edition (Dechema-Werkstoff-Tabelle). The Dechema-Literatur-Schnelldienst (Dechema Review and Reporting Service) also forms a part of the activities of the Dechema Institute. The demonstration and display halls of the Institute will enable visitors, both German and foreign, to obtain a broad birds-eye view of the present position of the chemical apparatus and equipment field. At the same time, visitors will also be afforded an opportunity of operating and testing the equipment and apparatus themselves. It is also intended that discussion groups shall be formed in connection with the demonstrations. Plans are also being formulated whereby German foreign students will be able to take part in a series of lectures and educational group meetings to be held during the University vacations.

* See P. Grodzinski—"The Meaning of Boart," *Ind. Diamond Rev.* 1952, 12, 233.

Faults in Pressure Die Castings—I

By W. M. Halliday

The factors influencing the quality of pressure die castings are many and varied, ranging from die design to the metallurgical characteristics of the alloy used. The main types of fault include dimensional inaccuracy, poor surface and mechanical weakness. In the present article, the author follows a general introduction with a more detailed treatment of faults in the first category.

DURING recent years, considerable attention has been devoted to the development and improvement of the pressure die casting process, and a good deal of highly informative and valuable practical information has been published on several important aspects of the subject. In the main, this has been concerned with the following: (a) the evolution of more successful component and die designs, conforming ever more closely to the inherent requirements and exigencies of the process; (b) improvements in die construction methods and materials; (c) greater knowledge of the metallurgical and physical properties of die casting alloys and their behaviour on the foundry floor; and (d) the more economical manipulation of such alloys in modern die casting machines. In contrast to this valuable work, comparatively little attention has been devoted to the equally important subject of component defects and inaccuracies—their cause, effect, diagnosis, and cure. Little technical data has, in fact, been published on this aspect of die casting.

Because of the high speed of casting production now possible, especially when employing multiple impression dies, the rapid detection and correction of any defects or inaccuracies is, of course, of the utmost practical importance. Inability to accomplish this may lead to the production of large quantities of unsatisfactory castings, which may be rejected by the customer, or cause a large amount of corrective work in the die casting foundry; it may also lead to speedier failure of the die. Success in this matter can only be attained if the responsible die casting engineer possesses an extensive practical experience of all the vagaries of the process—the behaviour of alloys—and the methods to be adopted for the detection of faults, the diagnosis of their origin, and the subsequent modifications necessary for their elimination.

Contrary to what might be expected of what is inherently a metallurgical casting process, the number of faults in components is quite small, especially in so far as these may constitute reasons for scrapping castings. On the other hand, it should be pointed out that the causes of such inaccuracies and defects may be extremely numerous and varied. Some of these causes are simple in character, and relatively easy to determine and correct. Others, however, may be very complex in nature, deep seated in origin, difficult to isolate, and far from easy to eliminate.

Factors Controlling Quality

The character, frequency of occurrence, and seriousness of such faults are usually conditioned by the general accuracy and quality of the die; the suitability of the design features embodied in that tool; the type of alloy; and the whole casting technique followed.

With all die casting operations, there are three basic elements determining the quality of the finished die cast products. When puzzling and seemingly inexplicable faults arise, each element has to be carefully analysed and checked in a systematic manner in any effort to locate the underlying causes.

These three conditioning elements are:—

- (1) The die casting die.
- (2) The die casting alloy.
- (3) The die casting machine.

The first element includes not only the fundamental design and construction features of the tool, together with their suitability for economically producing the casting required, but also the manner in which the die is to be operated, i.e., its working temperature, cavity lay-outs, general accuracy, coring, ejection and so forth.

The last of the listed elements includes both the manner in which the die casting machine is to be manipulated, and the whole set of sequential movements appertaining to both machine, die, and alloy charge. Each die is generally a unique kind of mechanism calling for a special mode of operation, and having an equally unique set of manipulation problems. Such features and problems generally concern the gating provisions; the type of ejection; the speed at which such ejection can be accomplished; the working temperature; the manner in which the last-named is maintained at the correct level throughout long periods of casting; and so forth.

Most dimensional inaccuracies and variations result from errors in the critical cavity configurations in the die, or from discrepancies in the working of that tool. Unequalised expansion and contraction of vital die elements, or the alloy charge, may also produce the same component errors. Likewise, the surface blemishes present on the casting, may be faithful reproductions of faults present on the walls of the cavity formations in the die, or may result from the operation of the tool at an incorrect temperature, etc. Some of the physical irregularities and weaknesses may also originate from defects in the die design or construction.

Modern die casting alloys are now available in such a reliable and stable form that little trouble need be experienced in their use, providing, of course, that any subsequent alloying carried out in the foundry is performed with the utmost care, so as to safeguard the initial purity of the metals. Stringent precautions are desirable in this respect to avoid accidental contamination of the resultant alloy.

General Classification of Faults

The types of faults encountered in die cast articles may broadly be classified as follows:—

- (1) Inaccurate dimensions of essential parts of the casting. Such errors may be exhibited as variations

on certain dimensions, as excessive ovality, as non-parallelism, or as a variation in the degree of uniformity of both shape and size occurring with successively cast components.

(2) Faults detracting from the quality of the surface finish and the general appearance of the casting.

(3) Faults entailing a considerable weakening of the mechanical strength, and/or a loss of certain physical characteristics or properties.

Incidentally, it is sometimes assumed that *all dimensions* on a pressure die casting can be consistently held to the same close limits over large batches. This is rarely the case in practice, however, because of several factors. For example, the disposition of the parting planes in the die has an important bearing on the precision obtained with certain component dimensions. If the parting line has to be situated across the diameter of critically toleranced portions, some ovality may be expected. Those portions of the casting which have to be produced with fine accuracy should, therefore, preferably be situated wholly in one die block.

Should the die be of built-up construction, the location of the various intersecting jointing planes and surfaces may have a considerable bearing upon the closeness to drawing tolerances attained with certain dimensions. Often such inaccuracies may be avoided by using a solid-block type of die construction devoid of such joints.

The disposition of cavities and cores, and particularly the manner in which movable cores are mounted in slides or other bearing mechanisms, may likewise affect the degree of accuracy in the finished casting. To a lesser extent, the manner in which the die is gated; the relationship of cavities to runner channels and the sprue; the situation of ejector rods; and the points at which these impinge on the casting surface; also influence accuracy.

Usually the detection and correction of dimensional errors arising from the above causes is comparatively simple. On the other hand, in certain cases, the main cause may lie in the selection of an unsuitable design of die or component. For instance, the component may have very heavy and thick wall sections joined directly to slender thin walls. Again, cores may be situated in the slender portions, or certain parts of the casting wall may be cored out to form long unsupported sections, and so forth.

Component designs of this nature usually present some difficulties, especially if precision tolerances have to be maintained, and if the designer is not permitted to employ strengthening features in the shape of ribs, beadings, bosses and fillets. Dimensional variations or errors then arise because of the effects of excessive distortion and warpage of the component. Cracking of such slender portions may also be expected in such circumstances.

Surface defects and irregularities of the second class may result from the inferior quality of the finish imparted to the cavity walls, etc. They may also be caused by the use of contaminated alloy. Overheating of the die or alloy charge is also a frequent source of serious surface blemishes. Then again, the use of an unsuitable die lubricant or wash may cause certain forms of surface defect to appear on castings. Another cause of such faults is the failure to close the die blocks properly when injecting the charge. This will result in a thickened parting-line flash, which, upon removal, will leave an unsightly scar or marking on the casting.

Component defects listed under the third heading generally prove the most serious and difficult to correct. They may arise from a variety of causes, chief of which are as follows: (a) unsuitable die or component design forms, entailing unduly great variations in wall thickness of the part; (b) deep or large cored holes, especially where the core plugs cannot be provided with ample relief or draft to simplify extraction of the component; (c) wrong type of inserts may be used; and (d) there may be poor lay-out of the cavity formations, resulting in excessively long runner flow, and consequent over-rapid chilling of the metal stream before it reaches the die cavity.

Weakening defects may also result from insufficient injection pressure or speed, or the alloy itself may be contaminated to such a harmful extent as to lead to the fracture of slender walls when subjected to loading stresses. Air may be trapped in the die cavity by the incoming alloy stream due to the die being improperly air vented. On the other hand, gas or fume may be created in the die cavity as a result of the burning of excess lubricants, or certain substances contained in the die coating materials. Again, there may be excessive turbulence in the die cavity, due to wrong location of the gate and poor flowing facilities thereby afforded the alloy charge. Finally, the die or metal charge may be grossly overheated, thus requiring a much longer time to solidify before the die can be parted for the ejection of the casting.

Considering each of the foregoing three general classes of die casting faults in closer detail, the following individual defects and inaccuracies are worthy of particular note.

Dimensional Inaccuracies

These may arise because of any, singly or in combination, of the following causes.

(a) *Lack of accuracy on vital dimensions in the die cavity.*

(b) *The presence of excessive working clearances between critical moving parts of the die mechanism, due to lack of precision when initially manufacturing the die, or to onset of wear or damage after use.*

(c) *The thermal instability of the die when in operation.* Failure to ensure a uniform heat distribution, or the correct temperature level over the vital cavity area, etc., may result in critical dimensions on the casting varying an undue amount, due to uneven rates of shrinkage. Two distinct forms of shrinkage and expansion have to be considered in this connection. Firstly, there is die contraction, and, secondly, the natural shrinkage of the solidifying alloy. Incidentally, such variations may occur whilst the casting is still in the die, or at a later stage when it has been ejected therefrom and allowed to cool to room temperature.

(d) *The degree of accuracy attained between essential working elements of the die mechanism.* This cause is usually absent in a completely new die, but tends progressively to develop in seriousness as wear occurs between die members.

(e) *The manner of ejection of the casting from the die.* The ejector rods must be located judiciously in relationship to cores, slender or unsupported wall sections of the casting, runners and gates, etc.

Where such provisions are of an unsuitable character, the casting may be excessively distorted during the

actual ejection stage, to such an extent that critical finely tolerated dimensions cannot be held to the specified close limits.

The interval elapsing after injection of the charge prior to ejection of the casting is also of considerable importance when producing precision type castings. If the component is extensively cored out, or has large surface area and great length in relation to wall thickness, the time allowed before ejecting has to be carefully determined, by test and try methods performed on the actual die under normal working conditions.

By leaving the casting over-long in the die cavity, it may be found that distortion occurs in certain areas, or in respect of cored holes, due to the inability of the casting to contract freely in all directions, because of the interposition of cores, etc. Slender wall sections may thus be over-stressed and drawn out of correct shape and size, whilst important hole centres and diameters vary similarly.

On the other hand, too rapid ejection of the casting may mean that certain wall portions become drastically warped from the following cause. The inner layers of the thick and heavily massed wall sections are in a molten or plastic condition much longer than those in the thin wall portions. Thus, unequalised contraction rates accrue, and the still shrinking heavy sections have to draw additional metal from some other portion of the casting. This material cannot be drawn from the already solidified slender wall sections, although greatly increased stressing occurs in these portions. The surface layers surrounding those contracting heavy sections thus tend to be drawn inwards, producing a "sink" or depressed surface area. Coupled with this condition, there is almost certain to be some loss of, or variations in, the size of the part at the affected points.

Unequalised shrinkage of the above kind also tends to cause some stretching of the adjacent thin wall sections, by reason of the excessive forces set up by the continued contraction of the heavy sections. Thus, again, variations in size result.

(f) *The location of core plugs, core-slides, ejector rods, inserts and similar mechanisms.* Generally speaking, the main ejecting pressure should be concentrated around, or as near as practicable to, the largest size of fixed core employed in the die. The rods should be distributed in such a way as to ensure equalised pressure being exerted on the walls of the casting. All ejector rods should preferably be anchored to the same plate, so that their movements commence and finish at exactly the same instant and their contact with the wall of the casting is uniform throughout the operation.

Core plugs should be provided with the maximum amount of side taper or draft to minimise ejection pressures. In some instances, it may prove better and safer to extract the casting by means of retractable cores instead of by ejector push rods. Drawing a core plug out of the die block, so stripping the casting therefrom, may result in a much more even load being imposed on the casting, thus reducing greatly the tendencies towards distortion, errors, and cracking.

Sometimes ejector rods may be incorrectly set in relation to the base or wall of the cavity into which they have to interject. If the end of such rods lie below the cavity wall, shallow bosses are formed on the surface of the casting. The presence of such projections has the effect of restricting normal shrinkage whilst the casting

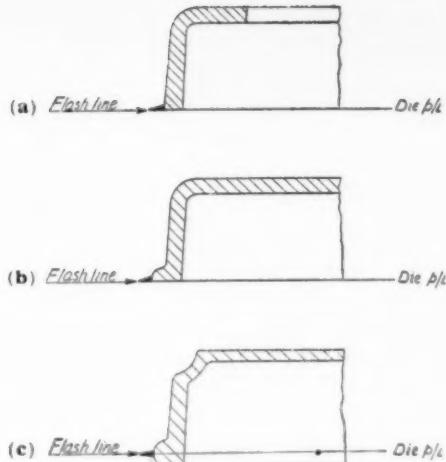


Fig. 1.—Use of beading to facilitate removal of parting-line flash.

is in the cavity, and again dimensional inaccuracies can be expected.

The same effect obtains if the rods are allowed to project too far forward into the cavity area. Additional strain arises when ejecting a casting formed under these latter conditions, which might be of such a magnitude as to cause dimensional errors.

(g) *Heavy flashing on the component*, due to improper sealing at parting-line surfaces, or jointing faces of the die blocks.

The importance of beadings; ejector-rod location; the situation and mounting of cores; and location of die parting joint, etc., are illustrated by the diagrams Figs. 1 to 5.

The three diagrams in Fig. 1 illustrate the advantages to be gained by the use of beadings on an external wall of a component to facilitate removal of parting-line flash. Fig. 1a shows a partial sectioned view of a hollow box casting, where the main die parting lies across the bottom (open) end, and where the side walls, both externally and internally, have but slight inclination to the parting-line joint.

With this first example, the side walls are perfectly straight right up to the parting line, and are devoid of any beading, etc., at the outer edge thereof. The heavy black shading shows where thick flash normally develops on the side of the casting, due to a momentary partial opening of the die blocks during injection, and/or as a result of distortion or wear of the parting-line surfaces of the die. The intervention of any other feature liable to impair close sealing of the die blocks, also results in flash.

Removal of flash from this edge involves certain difficulties, whether it be achieved by pressing the casting through the usual trimming die; by finishing the flat (open) end; or by manual means with file or scraping tool. The first and third methods entail a risk of shearing, gouging, or scoring the surfaces on the lower part of the walls, whilst the second method requires the removal of a considerable amount of metal from the part.

Fig. 1b shows a much preferred design, employing a raised beading on the outside of the wall. This beading

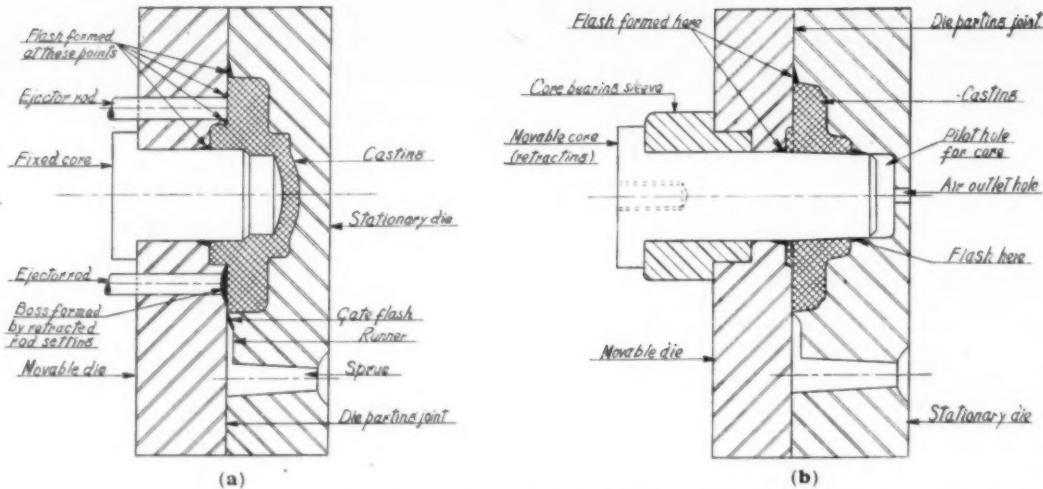


Fig. 2.—The development of ejector rod flash and its limitation.

is a quarter circle in section, its flat top being exactly in line with the parting-line surface of the component. With a casting beaded in this manner, parting-line flash occurs at the crest of the beading (see heavy black shading), and its removal is much simpler and safer than with the preceding design, whether the customary trimming die, or hand file methods are used. It is only necessary to adapt the cavity formation in a single die block to reproduce such a beading, thus making for simplified tooling and low cost.

Fig 1c shows another common type of beading often employed to expedite de-flashing, among other reasons. This is semi-circular in section, and is reproduced from a cavity adaptation formed in *both* die blocks. Thus, it is more difficult to reproduce with close accuracy and neatness than that shown in Fig 1b. The die parting line passes through the maximum diameter of the bead, and flash occurs on the crest, as shown in solid black shading.

An exact matching operation arises when machining the cavity formations in the die halves, and subsequent maintenance of correct alignments of the bead portions will depend greatly upon absence of wear, distortion, or deflection of dowels. Flash is removable with the same ease and safety as with the previous example.

The development of ejector-rod flash is illustrated by the diagrams in Fig. 2. The left-hand diagram (Fig 2a) shows the essential die construction for a hollow cap casting, which has a large two diameter cored blind hole and a concentric circular spigot formed in the movable die block. The remainder of the cap casting is reproduced by a large circular cavity in the stationary die block.

The core plug is fixedly mounted in the movable die, and the casting is ejected by four rods sliding in that block. These rods bear against the left hand side of the part, and flash may occur at the points indicated in solid black shading.

Removal of the collar flash around the ejector-rod contact points cannot easily be effected, except by lightly turning the end face of the part, thereby destroying, perhaps, some critical dimension.

The right-hand diagram (Fig 2b) shows a very useful and economical alternative die design which avoids the above difficulties, and, by eliminating the need for

ejector rods, reduces the number of possible flash points on the casting.

In this example, the component is very similar to that in Fig 2a, except in so far as the centre hole is cored fully through. The core, mounted in the movable die block, is thus guided in a hole in the stationary block beyond the floor of the cavity. Additional guidance is obtained for the core plug by a bearing sleeve fixed within the moving block, which latter may thus be maintained to small thickness.

After injection and opening of the die, the casting adheres to the core in the movable die half. At a certain stage, the movement of the latter is arrested, the core plug then being partially retracted into the die a distance sufficient to strip the casting off its front end. Thus, no ejector-rods are required. This design is extremely useful as a means of limiting flash formation on critical surfaces.

Fig. 3 shows a common coring requirement which is prone to lead to excessive flashing on the part. This sectioned illustration depicts the die construction for a casting having a cored hexagon hole extending the full length. Tooling difficulties, precision matching, and high costs would arise if such core plug were to be hexagon in shape throughout its full length. This would necessitate guide and pilot holes of similar hexagon shape in the movable and stationary die blocks. The core would have to fit extremely closely on all sides in such holes, and any interstices greater than about 0.002 in. would result in flashing at the end of the hexagon hole. Considerable binding might also occur between the core and the pilot hole in the stationary die.

A much preferred design is illustrated in Fig. 3. The core is formed with a large *cylindrical* shank for fitting into the movable die, and a smaller round pilot stalk engaging in the stationary die. The length of the hexagon portion in the middle of the core is equal to the thickness of the casting. A very close seal must be ensured between the right hand end of this hexagon portion and the floor of the cavity in the stationary die, when the blocks are closed for injection as shown here.

With this construction, flash can be expected at points indicated by solid shading, particularly heavy flash occurring at the right hand side of the part around

the edges of the hole, should the die blocks fail to close properly.

Thick flash at this point can easily be avoided by shortening the length of the cored hole so that the cylindrical pilot stalk of the core extends slightly into the casting, as shown by heavy broken lines. The core does not then have to seal closely against the base of the cavity in the stationary die, and improper closure of die blocks does not cause heavy flash as with the previous design.

Fig. 4 illustrates the relationship between component shape and die parting joint, and how the former may be modified to eliminate flashings on critical surfaces due to the presence of the die parting.

The upper diagram, Fig 4a, shows a simple lever component, requiring a stepped parting joint in the die, owing to the end of the arm portion being semi-circular in section.

To make a casting this shape, the die has to be parted in a double stepped manner as shown. Flashing can occur across each side of the arm portion, and around the middle of the semi-circular beading at the base of the large boss on the left. Removal of this flash would be difficult and costly, and may easily result in unsightly scars on the sides of the component.

Fig. 4b shows a modified component design permitting a simple straight parting joint lying wholly in one plane in the die. The end of the arm portion is rounded quarter circle at the top corner only, a square corner being left at the bottom edge. The beading is also made quarter circle in section, with its crest coinciding with the bottom of the arm portion.

Flash with such a simple die joint occurs only on the straight base of the arm and around the crest of the beading, which can easily be trimmed by pressing the part through a die, without incurring damage or scoring the sides of the arm, etc.

It is often possible, by modifications to component designs, to ensure simpler forms of die parting, reduced flashing, and elimination of marks and flashes from essential polished surfaces. Fig. 5 depicts the influence of die parting joint location and flashing effects with a component having a deep undercut circumferentially around the exterior.

This casting has an enlarged circular flange, a domed head, and a relatively narrow deep undercut between

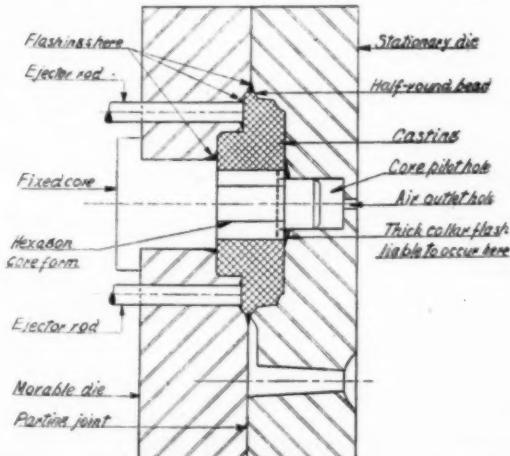


Fig. 3.—Flash on casting with hexagonal cored hole.

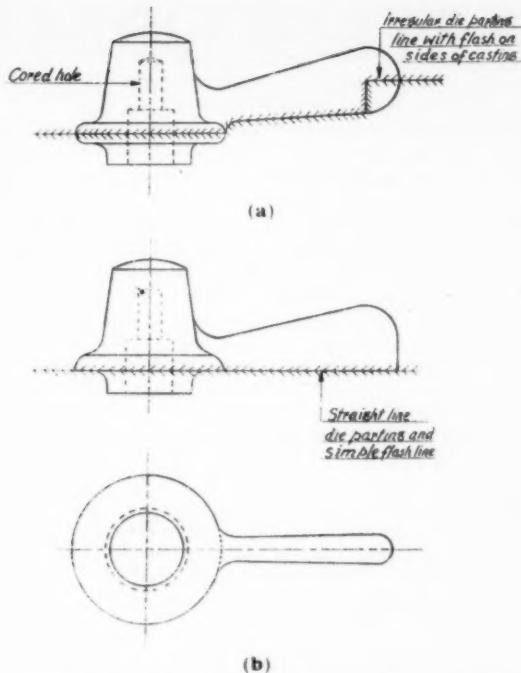


Fig. 4.—Relationship between component shape and die parting joint.

these two portions, the root diameter of the undercut being considerably less than that of the domed head or flange. The end of the parallel cylindrical shank at the opposite side of the flange is cored hollow for a certain distance.

Two alternative die constructions are available to produce this component. That seen in Fig. 5a shows a die construction in which the parting line runs transversely around the component as indicated by line X-X. With this die, a cavity formation exactly half the component shape is provided in each die block. The movable die has a retracting slide working in the vertical plane, and passing well beyond the parting line into a guideway in the opposite die block. This slide carries the core plugs for hollowing out the component shank.

It is possible to employ several cavities in such a die, these being situated side by side, so that all cores are mounted in a single slide. The slide may be actuated by the usual inclined pins affixed in the stationary die block.

The cavity is gated at the bottom of the domed head, as shown, and ejection is effected by means of the customary ejector-rods mounted horizontally in the movable die. One rod bears on the upper shank portion of the casting, the lower rod bearing on the runner material slightly below the actual gate.

With this construction, it will be understood that some flash occurs transversely around the middle of the part, coincident with the die parting joint. In addition, the severed gate on the domed head constitutes additional flash, whilst the upper ejector-rod forms collar flash on the side of the shank. Removal of flash of this extensive character is both difficult and costly, especially in respect of the deep undercut portion.

With this particular component, the domed head had to be highly polished and chromium plated to a bright

finish, hence scars and marks were undesirable. Fig. 5b shows the second die construction, which is preferable because the extent of flashing is very much reduced, and, moreover, the domed head is reproduced intact, since the gate is located on another portion of the casting.

With this construction, the main parting joint is located around the largest diameter end of the flange (see line Y-Y). An additional flash line lies at the root of the undercut portion at each side, as shown coinciding with the sealing faces of two opposed slides in the die. No flash occurs on the domed head.

The major portion of the cavity is contained in the stationary die block, which also carries two slides, located one above and one below the cavity and acting in opposite directions. These slides have narrow extensions passing into the cavity area, being shaped to form one half of the undercut, in each case. The slides are operated by external means before opening the die blocks. Fixed cores are mounted in the movable block for hollowing the end of the shank, and ejection is effected by means of four small rods, working horizontally in that die half to bear on the end face of the flange of the casting. The gate is situated at the periphery of the largest diameter portion of the flange as shown.

Collar flash is liable to occur at the points of contact between ejector rods and casting, but these can easily be removed from the flat side face. With this latter die construction, not only are flashing tendencies greatly minimised, but it is also possible to maintain closer control over the various diameters of the component, since none of these, with the exception of the root of the narrow undercut, are intersected by a die parting joint.

Such a die form is, of course, more costly to produce, and possibly slower to operate than the preceding example, but the better surface appearance, reduced flash, and greater precision on critical dimensions should amply recompense for such extra constructional expense.

The cure for practically all dimensional inaccuracies lies in simple corrections to the die. Most of these present no great difficulties and, of course, much may be accomplished to counteract their development by ensuring that the die is used with the greatest care. For example, when castings are found to stick in the die cavity, operators should be forbidden to use steel pry bars to eject or ease them.

Serious damage to the smoothly polished walls of the die cavity or cores may arise during periods of storage when the die is not in production use. The inner surfaces of the die should be cleaned and lubricated, or given a coating of anti-corrosion fluid, and the die blocks placed in their closed position, the whole tool being suitably packed against ingress of moisture, dirt, fume, or contact with other metallic articles. The greatest care

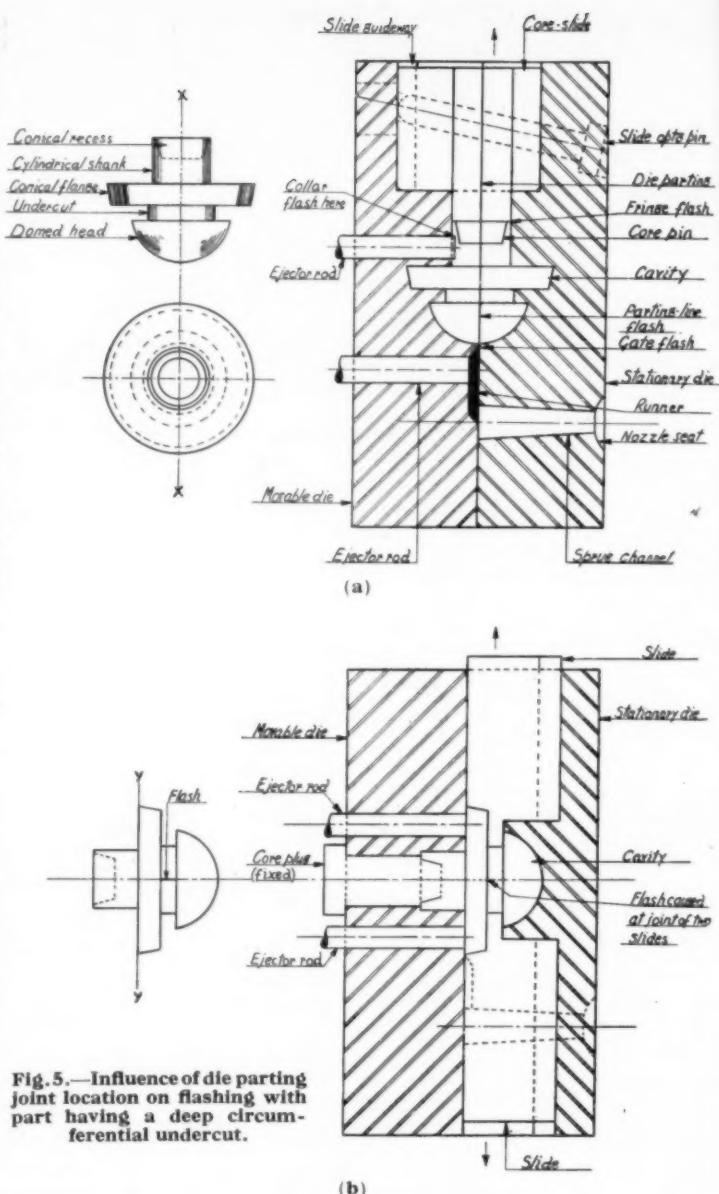


Fig. 5.—Influence of die parting joint location on flashing with part having a deep circumferential undercut.

should always be taken of all dies taken out of production and proper storage facilities provided.

Dollar Order for English Electric

THE contract for the manufacture and installation of two water wheel generators, physically ranking among the largest ever built, has been awarded to the English Electric Export and Trading Co. Ltd., by the United States Army Corps of Engineers at a cost of \$3,651,476 (approx. £1,300,000) which was the lowest bid received. The maximum output of each generator will be 84,740 kVA at 13.8 kV, 60 cycles and 85.7 r.p.m. Well over 1,000 tons of material, mainly special steels and copper, will be required for each unit, which will have an overall size of nearly 52 ft.

NEWS AND ANNOUNCEMENTS

The Institute of Metals

Award of Medals

The Council of the Institute of Metals has made the following awards of medals for 1954 :

The Institute of Metals (Platinum) Medal to DR. LESLIE ARCHISON, in recognition of his services to metallurgy in industry, in education and in public service.

The Rosenhain Medal to Professor A. H. COTTRILL, (Professor of Physical Metallurgy, University of Birmingham) in recognition of his outstanding contributions to knowledge in the field of physical metallurgy, with special reference to the deformation of metals.

Atomic Energy Lectures

SPONSORED by the University of London Department of Extra-Mural Studies and The Atomic Scientists Association, a course of six lectures on Atomic Energy will be held in the Beveridge Hall, Senate House, Malet Street, London, W.C.1, on successive Wednesdays at 7 p.m., commencing 20th January, 1954. The subjects and speakers are as follows : January 20th, Atomic Research at Harwell (Sir John D. Cockcroft) ; January 27th, Atomic Weapons (Prof. O. R. Frisch) ; February 3rd, Power from Atomic Energy (Prof. F. E. Simon) ; February 10th, Radiation Hazards of Atomic Energy (Dr. J. F. Loutit) ; February 17th, Medical Uses of Atomic Energy (Dr. E. E. Pochin) ; and February 24th, Atomic Energy and Moral Issues (Prof. Kathleen Lonsdale and Sir George P. Thomson).

The admission fee for the course of six lectures is 10s. (Members of Atomic Scientists Association, 5s.) and for single lectures, 2s. Applications for tickets for admission to the course should be addressed to the Cashier, University of London, Senate House, Malet Street, W.C.1, and marked "Extension Courses" on the cover. Cheques, money orders and postal orders should be made payable to the University of London and crossed Westminster Bank, Tavistock Square Branch, University of London A/c. Tickets may also be obtained at the lecture room on the occasion of each lecture.

The Physical Society Exhibition of Scientific Instruments and Apparatus

THE 38th (1954) Annual Exhibition of the Physical Society will be held at the Imperial College of Science and Technology, Imperial Institute Road, London, S.W.7, from Thursday, 8th, to Tuesday, 13th April, 1954. The Exhibition will be wholly located in the Physics and Chemistry Departments of the College and this should be of great convenience to visitors. Arrangements for issuing tickets have been greatly simplified and applications for tickets should be made to the Secretary-Editor at the offices of the Society, 1, Lowther Gardens, Prince Consort Road, London, S.W.7. Tickets will not be sent out until the beginning of March, and no order can be acknowledged unless a stamped addressed post card is also enclosed.

As in previous years the comprehensive Handbook of the Exhibition will be available at the Exhibition and copies can be obtained on application to the Secretary-Editor at the address given above. The price of the publication is 6s. (by post 7s. 3d.).

Mond Nickel Fellowship Awards

THE MOND NICKEL FELLOWSHIPS COMMITTEE announces the following awards for 1953 :—

J. E. BENSON—(Metropolitan Vickers Electrical Co., Ltd., Manchester) to study the technique and interpretation of results of non-destructive testing of metal components in the United Kingdom, on the Continent and in the U.S.A. and Canada.

K. BLACKBURN—(Dorman, Long & Co., Ltd., Redcar) to study hot-metal basic open-hearth practice in Great Britain, on the Continent and in the U.S.A. and Canada, with particular reference to mixer furnace operation, refractories, instrumentation and pitside practice.

N. B. PRATT—(Broken Hill Proprietary Co., Ltd., Newcastle, N.S.W.) to study the technical and economic aspects of recent advances in the erection and operation of integrated iron and steel works in Great Britain, on the Continent and in the U.S.A. and Canada.

I.M.M. Travelling Fellowships

A LIMITED number of Travelling Fellowships will be offered by the Institution of Mining and Metallurgy to members of the teaching staff of universities and approved schools of mining and metallurgy in the United Kingdom to enable them to visit important mining and metallurgical centres and research establishments in the 1954 long vacation. Applicants should be 30 years of age or over, and should be engaged in the teaching of mining engineering, extraction metallurgy, or economic geology. The tenure and value of each Fellowship will be decided according to individual requirements, but in general the tenure will not exceed three months and the value will not be greater than £500. The award is intended to meet the travelling expenses of the Fellow.

Forms of application for these Fellowships may be obtained from the Secretary, The Institution of Mining and Metallurgy, Salisbury House, Finsbury Circus, London, E.C.2, to whom they should be returned after completion *not later than 1st March, 1954*. It is hoped that it may be possible to make similar awards in 1955.

International Electron Microscopy Conference

THE Joint Commission on Electron Microscopy is sponsoring an International Conference in its subject : at the invitation of the Electron Microscopy Group of the Institute of Physics, this will be held in London from July 16th-21st, 1954. Sessions will be devoted to the following subjects : (1) instruments and associated apparatus ; (2) electron optics ; (3) attainment of very high resolution ; (4) reflection methods ; (5) metallurgical applications ; (6) industrial and chemical applications ; (7) fixation and drying methods ; (8) microanatomy of cilia, etc. ; (9) fibrillar structures ; (10) Structure of cell walls ; (11) internal structure of cells ; (12) bacteria ; (13) viruses ; (14) biological specimen techniques ; (15) metallurgical and general specimen techniques ; and (16) effect of electron irradiation on matter. A number of survey lectures are being invited.

Offers of papers for consideration by the Committee are cordially invited, and should be notified on a special form as soon as possible, and not later than March 1st.

An abstract of the proposed paper (in English, French or German), not exceeding 50 words should be included. Acceptances will be notified soon after that date and more detailed abstracts submitted by the date of the Conference.

Further particulars regarding papers and enrolment can be obtained from the Conference Joint Secretaries ; Dr. J. E. Challice and Mr. F. W. Cuckow, c/o, Institute of Physics, 47, Belgrave Square, London, S.W.1.

1954 Mond Nickel Fellowships

THE Mond Nickel Fellowships Committee now invites applications for the award of Mond Nickel Fellowships for 1954. The main object of these Fellowships is to enable selected applicants of British Nationality and educated to University degree or equivalent standard to obtain additional training and wider experience in industrial establishments, at home or abroad, so that, if they are subsequently employed in executive or administrative positions in the British metallurgical industries, they will be better qualified to appreciate the technological significance of research and to apply its results.

There are no age limits, though awards will seldom be made to persons over 35 years of age. Each Fellowship will occupy one full working year. It is hoped to award five Fellowships each year, of an approximate value of £900 to £1,200 each. Applicants will be required to define the programme of training in respect of which they are applying for an award, as well as particulars of their education, qualifications and previous career. Full particulars and forms of application can be obtained from : The Secretary, Mond Nickel Fellowships Committee, 4, Grosvenor Gardens, London, S.W.1. Completed application forms will be required to reach the Secretary of the Committee not later than June 1st, 1954.

Lectures on Rarer Base Metals

A COURSE of eleven lectures on The Rarer Base Metals—Production and Modern Applications, will be held in the Department of Chemistry and Biology of Acton Technical College at 7.30 p.m. on successive Friday evenings, commencing January 15th, 1954. The subjects to be discussed in chronological order are (a) The Rarer Base Metals and the Periodic System ; (b) Beryllium and its Applications ; (c) Gallium, Indium and Thallium ; (d) Titanium and Zirconium (two lectures) ; (e) Germanium and Silicon and their uses as Semi-Conductors ; (f) Vanadium and its Applications ; (g) Niobium and Tantalum (two lectures) ; (h) Molybdenum and Tungsten (two lectures). The fee for the course is 30s.

Further particulars can be obtained from the Principal, Acton Technical College, High Street, Acton, W.3.

Turkish Contract for Brightside

A CONTRACT of well over £500,000 for the supply of a blooming and slabbing mill for the Karabuk Steel Works, Karabuk, Turkey, has been secured by The Brightside Foundry and Engineering Co., Ltd., Sheffield, against keen competition from Germany. In announcing the contract at the Company's staff dance, the Managing Director, Mr. T. C. Firth, said that the negotiations were made more difficult by the recent disturbance in the engineering industry. To obtain export business, it was essential to create confidence in the ability of this country to produce the highest class of product at the right price and at the right time.

Conference on the Physics of Particle Size Analysis

AT The Institute of Physics' Conference on "The Physics of Particle Size Analysis," which is to be held in Nottingham from 6th-9th April, 1954, the following sessions have been arranged : (1) the motion of particles in fluids ; (2) the scattering of light by particles ; (3) the general phenomena encountered in particle size analysis ; and (4) the comparison of methods and the automated methods of particle counting and sizing. Preprints of the papers to be presented to the Conference will be available beforehand and, together with a summary of the discussion, will, in due course, be issued as a supplement to the *British Journal of Applied Physics*. Further particulars may be obtained from the Secretary, The Institute of Physics, 47, Belgrave Square, London, S.W.1.

Darwin's Group Acquisition

FOLLOWING the repeal of the Iron and Steel Act, 1949, and the dissolution of the Iron and Steel Corporation of Great Britain, the Darwins Group Board of Directors announces the acquisition of the whole of the share capital of the old established and well-known Company, The Sheffield Forge and Rolling Mills Co., Ltd. This company will continue to trade under its usual title, but will have at its disposal the additional and comprehensive facilities of the Darwins Group ; conversely the Darwins Group will benefit with the extended range of rolling sizes, etc.

A further announcement concerns the installation of a new basic electric arc furnace at Andrews Toledo, Ltd. In putting the new furnace into operation, the present capacity, maintained by acid open-hearth and high frequency induction furnaces, will be expanded so that the Group will be in a position to encompass within its range practically every steel in general usage.

Boron Supplies

THERE has been considerable interest recently in the improved properties of steels and other alloys containing boron. Many metallurgists do not seem to have realised that, even prior to the war, there were available in this country suitable boron addition alloys for iron and steel, the non-ferrous copper and nickel-base alloys, and the light metal industry, as well as smaller supplies of elemental boron. These have been produced for a number of years for direct metallurgical use as additions, and for special welding electrodes, by the Murex organisation, and much of the useful development work with boron has been based on these materials.

Standard Reference Block for Ultrasonic Testing

THE F.E.18 Ultrasonics Panel of the British Welding Research Association, is considering the need for a standard, single-hole, steel reference block for use in ultrasonic testing. Such blocks would need to be commercially available and should bear a mark showing that they are of approved design and quality. All firms and individuals interested in this possibility are asked to communicate with the Secretary, the F.E.18 Committee, British Welding Research Association, Abington Hall, Abington Cambridge, from whom further information may be obtained.

RECENT DEVELOPMENTS

MATERIALS : PROCESSES : EQUIPMENT

Deoxo Indicator

A RECENTLY developed instrument capable of accurate and continuous measurement of small quantities of oxygen or hydrogen in mixtures of these two gases, or separately in other "inert" gases, is based on the measurement of the heat of combination of oxygen and hydrogen. It employs a precious metal catalyst similar to that used in the well-known Deoxo purifiers and, therefore, it is necessary when measuring oxygen to have at least a slight excess of hydrogen (or of oxygen if hydrogen is being measured). The catalyst is contained in a specially constructed vacuum-jacketed calorimeter, and a multi-junction thermocouple measures the difference in temperature between the incoming and outgoing gas. Thus, an e.m.f. is produced, which enables continuous indication and, if required, a record to be obtained. Although all the gas being measured is combined on the catalyst, the indicated temperature rise is significantly below the theoretical figure. This is accounted for by heat losses; but, by controlling the flow at a predetermined figure, reliable reproducible results are obtained. To calibrate the instrument, the inlet gas is first passed through a Deoxo purifier (built in the instrument for the purpose), and then, known quantities of oxygen and/or hydrogen are added by means of a small electrolytic cell, before passing to the calorimeter. The electrical output of the instrument is directly proportional to the oxygen or hydrogen content of the gas, i.e., the calibration curve is linear, and, therefore, calibration at one point only is required. The temperature yield, i.e., the indicated temperature rise compared with the theoretical, is of the order of 65%, but the actual figure is not important so long as it remains substantially constant. For calibration it is, of course, necessary to control the gas flow rate accurately; but, in use, a variation of up to $\pm 10\%$ does not affect the reading significantly. A sensitivity adjustment enables

the indicator reading to correspond directly with the oxygen or hydrogen content. The instrument covers the range 0.01% up to 1% oxygen with good accuracy, and can, with decreasing accuracy, be made to indicate down to as low as 0.001%.

The working of the apparatus would be affected by the usual "poisons" which inactivate the Deoxo catalyst, and, therefore, the instrument includes scrubbing and drying bottles. Carbon monoxide inhibits the catalyst and, since it cannot readily be removed from the gas stream, the use of the indicator is not recommended when carbon monoxide is present. However, this gas is an inhibitor rather than a poison, in so far as its effect is not permanent; i.e., the catalyst soon recovers its full activity on passing CO-free gas. The indicator is not suitable for gases containing unsaturated hydrocarbon, as these compounds become hydrogenated in the presence of the catalyst, thus liberating heat and causing an erroneous indication.

Baker Platinum Ltd., 52, High Holborn, London, W.C.1.

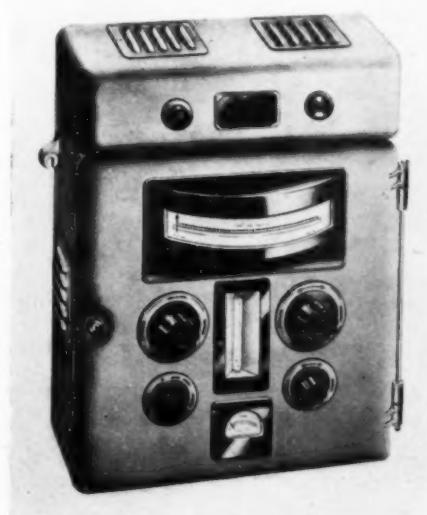
A High-Tensile Weldable Steel

THE search for new and improved steels occupies a large proportion of the research effort of metallurgists, and one objective over the years has been to provide a steel which has high strength and at the same time is readily weldable. The United Steel Companies Limited, as suppliers of steels covering a wide range of applications, regarded the development of steel of this type as a matter of considerable importance. Consequently, their Research and Development Department have examined many experimental steels only to find that additions of the more commonly used alloys produced steels which were generally difficult to weld when combined with satisfactory mechanical properties.

Another problem, having no direct connection with strong weldable steel, was also being studied. In an endeavour to improve the steel used in power station steam pipes, experiments were made to determine the effects of boron on the properties of low-carbon molybdenum steel. It was discovered that when molybdenum and boron were present in certain proportions, the yield point of a low-carbon steel was doubled without any necessity for heat treatment. As little as an ounce of boron, when added to a ton of steel containing 0.40% of molybdenum, was sufficient to produce this pronounced and important improvement in strength.

This new steel has been given the name of "Fortiweld," a title derived from the properties of the steel, namely: tensile strength of 40 tons/sq. in. combined with ease of welding, the latter property being attributable to its low carbon content. In addition to its high strength at ordinary temperatures and easy weldability, Fortiweld possesses good properties at 450°-500° C., being capable of withstanding service stresses two to three times those permissible for mild steel.

The unique combination of properties provided by Fortiweld make it particularly attractive for highly stressed welded structures either at ordinary or elevated temperatures, and many important applications have



already been made, including the casings of aircraft jet engines, special tubing and bridge components.

The United Steel Companies, Ltd., Westbourne Road, Sheffield, 10.

Oil Filter

The Stellar Filter of the Paterson Engineering Co., Ltd., of London has been specially developed for the fine filtration of rolling mill oils used in the production of highly finished aluminium plate and stainless steel strip. The oil is continuously re-circulated and, by bypassing a suitable quantity through a system of these filters, the level of contamination in the main bulk of oil is kept to a very low figure.

The filter consists essentially of a cylindrical shell and outlet manifold constructed to take Stellar patented wire-wound elements or candles, which are fine strainers and are employed as the foundation for a deposited filter bed of filtering powder, known as Stellafilt which may be kieselguhr, graded for the particular purpose, or an admixture of kieselguhr with other materials. The element consists of a tube with equally spaced longitudinal ribs, having a screw thread cut along the entire length, thus providing on each rib a series of grooves in which the consecutive turns of the wire wound on the ribs is firmly located. The openings over the whole area of the element are of regular size, and the result is a rigid element with the shortest possible flow and the minimum gap. The regularity of the openings ensures an even coating of filter aid, with no break which would permit unfiltered liquid to bypass: absolutely uniform filtration is thus ensured.

As regards materials of construction, filter shells for general use with aqueous liquids are made of mild steel and lined when necessary with a hard plastic enamel. For foodstuffs and corrosive conditions, stainless steel is used. The filter elements for water, oils and non-corrosive liquids are made with extruded brass tubular

cores, electro-tinned and wound with either Monel or stainless steel wire. Where it is imperative that the residual liquid should be conserved, special arrangements can be offered according to the nature of the application—siphon type elements, a reversible filter on trunnions, or a separate small filter for dealing with the contents of the main filter.

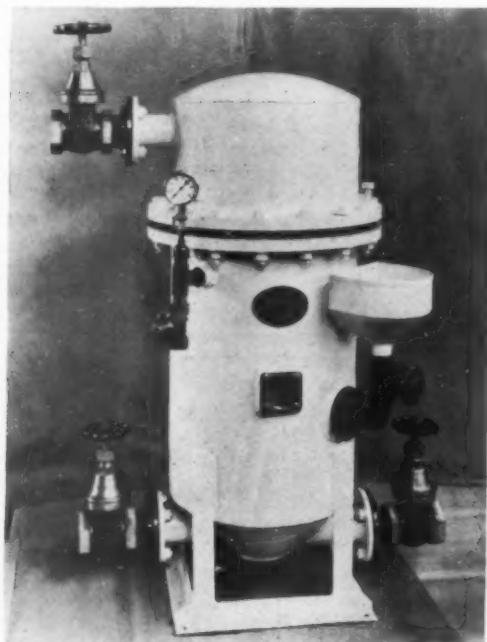
The filtration of liquids containing other than a small amount of suspended matter, is usually facilitated by the use of filter aid as a continuous addition throughout filtration. This addition of filter aid inhibits the formation of a skin of dirt on the pre-coat, and, by the deposition of further porous material, continuously maintains an open filter bed throughout the run. The Stellar Jet Filtraider enables this filter aid to be introduced proportionately into the stream of liquid immediately before the filter, and the abrasive diatomite does not, therefore, pass through the pump.

The Paterson Engineering Co., Ltd., London

A New Pirani Vacuum Gauge

THE development of a new Pirani vacuum gauge is announced by National Research Corporation, Massachusetts. Designed to operate in the pressure range between 1.0 and 0.001 mm. Hg (1000 to 1 microns), the Model 516 Pirani Vacuum Gauge has several unique interesting features. These include: (1) a low impedance power rectifier that eliminates the need for voltage adjustments during operation; (2) one meter that is used for both voltage adjustment and pressure reading; and (3) a compensating element mounted inside the cabinet; this facilitates installation and transportation of the gauge. There is, however, no loss of accuracy resulting from mounting the compensating head inside the cabinet. Another new feature is the incorporation of the outgassing circuitry on the pressure range selector switch. This permits outgassing at any pressure without danger to the filament. The gauge cord is 10 ft. long but cords up to 100 ft. in length may be used without loss of accuracy.

Pirani gauges are widely used in the electronics industry for manufacturing radio and TV tubes, in the plastics industry for pressure control in vacuum coaters, and in the metals industry on vacuum annealing and vacuum melting furnaces. Pirani gauges are used when fast and accurate readings are needed between 1000 and 1 micron pressures.



CURRENT LITERATURE

Book Notices

ENGINEERING DRAWING PRACTICE

1953 edition of British Standard No. 308. British Standards Institution, 2, Park Street, London, W.1. 10s. 6d.

ANY progress towards a universal drawing practice is of profound importance to engineers, and the publication by the British Standards Institution of a new and far more comprehensive edition of B.S.308—Engineering Drawing Practice—is undoubtedly a prominent milestone on the road towards such a world-wide understanding.

The need for a new standard was particularly felt during the second World War, when munition manufacture had to be sub-contracted to works of all kinds in all parts of the country, and indeed all over the world. It became more and more evident that the dimensioning and tolerancing of many engineering drawings, however well they conformed to accepted standards, did not always ensure that components would assemble or function correctly. The lack of precise definition of surface finish or geometrical accuracy may not be serious in some self-contained works producing specialised lines of manufacture, where the manufacturing technique, machine tool equipment, and inspection procedure have been built up by years of experience, and provided all essentials which are not clarified on the drawing. But where manufacture is sub-contracted, particularly on a widespread scale, there is no such background of experience to guide the unfortunate manufacturer or inspector, and no designer conveniently at hand to answer the questions which inevitably arise. Decisions become a matter of unguided opinion, and may result in a product not sufficiently accurate to function or even assemble properly, or, conversely, far more refined than is really necessary.

These difficulties led to the formation in 1944 of an Inter-Services Committee on Dimensioning and Tolerancing of Drawings. This Committee was responsible for the publication in 1948 of the manual "Dimensional Analysis of Engineering Designs," which laid down a number of fundamental principles which have, in general, been applied in the drafting of the new B.S.308. In the later stages of the work of the Inter-Services Committee several members of the B.S.I. Drawing Office Practice Committee were co-opted, and in 1949 a document on "Dimensioning and Tolerancing" was presented to the B.S.I. Committee with the suggestion that it should be widely circulated to obtain the views of engineers and, particularly, of industry.

This document was accordingly issued as a draft of B.S.308 Part II : 1949, and meetings were also held in various centres to amplify the new ideas embodied and to ensure the maximum of publicity. This draft, however, had a very cold reception. Some of the practices recommended aroused strong opposition from industry, and the manner in which the subject was presented did not commend itself to the majority. There was evidence, however, that even the strongest opponents recognised that most of the principles involved were basically sound.

It was, therefore, decided to make an entirely new approach to the subject, and a B.S.I. Sub-Committee was formed in September, 1950 to undertake a complete revision of B.S.308 : 1943, including a detailed

consideration of the dimensioning and tolerancing aspect. This drafting committee included representatives of the heavy electrical and mechanical engineering, aeronautical, automobile and radio industries, British Railways, the Ministry of Supply, the Admiralty and the National Physical Laboratory. It has held about 100 meetings during the past three years, and during this time has provided delegations to no less than seven international conferences on engineering drawing practice.

The new B.S.308 is divided into two sections. Section I (General Practice) is based on the previous B.S.308:1943, but while the section has been amplified in some respects, it has been considered advisable to confine the standard to recommended principles and methods to be followed in the preparation of engineering drawings. Subjects such as architectural drawings, survey plans, graphs, reproductions of drawings, and the nature and handling of drawing materials have been excluded, as such matters are more fully and appropriately dealt with in other British Standards. For these reasons it has been considered appropriate to alter the title of the standard from "Engineering Drawing Office Practice" to "Engineering Drawing Practice."

Section II (Dimensioning and Tolerancing) is entirely new, and presents a great advance in the technique of drawing statement. It lays down a number of general principles, of which the most important are that the drawing should define the finished product as required by the designer, and that dimensions which affect the function of the product should be expressed directly on the drawing. It also deals with ways of indicating dimensions, tolerances and notes on drawings, and with the dimensioning of common features, including a special section on the difficult subject of tapers.

The latter portion of the standard is devoted mainly to the principles and methods of expression of geometrical tolerances for straightness, flatness, parallelism, squareness, angularity, symmetry, concentricity, roundness and position. By the judicious application of these methods, it will be possible in future to define precisely acceptable limits of geometrical form, instead of relying on individual judgement or conjecture. It therefore rests with the design organization to decide how far it is necessary to apply geometrical tolerancing in each particular instance, having regard to functional requirements, interchangeability and probable manufacturing circumstances. The standard concludes with the dimensioning of profiles, and the indication of machining and surface finish.

INSTRUMENTATION AND CONTROL OF MILL FURNACES

By the Instruments Sub-Committee of the Plant Engineering Division of the British Iron and Steel Research Association. 34 pp., 8 illustrations. Obtainable from the Association, 11, Park Lane, London, W.1. 5s.

The selection, installation, maintenance and operation of instruments for rolling mill furnaces are dealt with by this booklet, which has been compiled by the Instruments Sub-Committee of B.I.S.R.A.'s Plant Engineering Division, under the Chairmanship of Mr. W. B. Wright of Colvilles, Ltd.

After an introduction enumerating the purposes and advantages of instrumentation the book is divided into four parts: (1) measurements and controls for operation

of mill furnaces; (2) examples of soaking pit and reheating furnace instrumentation systems at present in operation; (3) experiences and achievements in the use of automatic control; and (4) installation and maintenance of works instruments. An appendix tabulates instruments in common use, and their functions.

The basic measurements considered in Part I are air/fuel ratio, furnace pressure and steel temperature. Automatic control of both gas and oil fired furnaces are fully dealt with in the same part of the book. The examples of operating instrumentation systems given in Part II comprise new soaking pits by three different makers, respectively, at Colvilles, Ltd. (Clydebridge Works), Guest Keen Baldwins Iron and Steel Co., Ltd. (East Moors Works), Dorman, Long & Co., Ltd. (Cleveland Works), and a slab-reheating furnace by a fourth maker at the Abbey Works of the Steel Company of Wales, Ltd.

Among the instances of advantages accruing from automatic control given in Part III is that of a soaking pit which had to be put out of commission every six to eight weeks because of the formation of slag up to gas port level. Installation of automatic temperature and furnace pressure controllers enabled the furnace to operate for six monthly periods before it was necessary to remove the slag. This confined slag removal to the Christmas and summer repair periods, so that 100% availability was obtained.

New pits installed at the same plant, fitted with temperature and furnace pressure controllers, operated for fifty weeks before it became necessary to remove the slag. About 60 tons of slag were removed for a through-put of about 51,000 ingot tons. As the slag contained 78% Fe, this represented a steel loss of about 2 lb. per ton or 0.10%. At many pits this figure is about 0.4 cwt. to 0.5 cwt. per ton, representing about 2% loss. The authors point out that this reduction is almost certainly due to automatic control, because any breakdown of the controllers immediately results in an increased rate of slag formation.

The booklet contains eight line illustrations showing various instrument layouts and an instrument panel. Sir Charles Goodeve, F.R.S., contributes a foreword.

Books Received

"The Determination of Crystal Structures" (The Crystalline State—Vol. III). By H. Lipson, D.Sc., M.A., F.Inst.P., and W. Cochran, M.A. Ph.D. 345 pp. inc. numerous illustrations, references, name and subject indexes. London, 1953. G. Bell & Sons, Ltd. 50s. net.

"Select Methods of Metallurgical Analysis." By William Archibald Naish, Ph.D., A.R.S.M., B.Sc., F.R.I.C., John Edward Clennell, B.Sc., (London), Assoc. Inst.M.M., and Victor S. Kingswood, M.Sc., B.Sc., F.I.M. (Second Edition, Revised). 660 pp. inc. index. London, 1953. Chapman & Hall, Ltd. 75s. net.

"Engineering Metallurgy" (A textbook for users of metals). By Bradley Stoughton, Ph.B., B.S., D. Eng., Allison Butts, A.B., S.B., and Ardrey M. Bounds, B.S., M.S. (Fourth Edition). 479 pp. inc. index. New York, Toronto and London, 1953. McGraw-Hill Book Co., Inc. 53s. 6d.

"Applied Elasticity." By Chi-Teh Wang, Sc.D. 357 pp. inc. index. New York, Toronto and London, 1953. McGraw-Hill Book Co., Inc. 57s. 6d.

Trade Publications

We have received from Birlec, Ltd., Tyburn Road, Erdington, Birmingham, the first of a new series of catalogues and brochures which this Company now has available for distribution. These new publications are useful for reference purposes, for besides details of equipment, much useful metallurgical and technical information is included. In some sections analysis of comparative costs are included based on results obtained by users of Birlec equipment. Birlec, Ltd., are also finding that the closer attention now being paid to humidity control is focussing the attention of a variety of industries on the Birlec Lectrodryer air-conditioning equipment. A folder describing the recently introduced BE 40 unit designed for continuous operation in self-contained areas, such as storage rooms, is available. As a companion to this folder there is a pocket-sized card folder of Humidity Tables, showing the maximum amount of water vapour that can be present in air or other gases at any temperature before condensation occurs. Revised brochures dealing with: Shaker Hearth Conveyor Furnaces; Gas Carburising; Electric Furnace Brazing; and the Machine Tool Approach to Surface Hardening are available. There is also a new 20-page descriptive catalogue of Birlec Lectromelt Arc Furnaces and a publication describing, under the heading Machine Tool Approach to Production Heating, the induction heating equipment developed by Birlec, Ltd. These publications are freely available on application to the Company.

As a convenient means for checking the composition of metals and alloys, quantitative spectrochemical methods of analysis are widely adopted, and for several years there has been a demand for reliable standards, suitable for these methods, that can readily be obtained when required by prospective users who lack the facilities or the time to prepare their own standards. Johnson, Matthey & Co., Ltd., whose existing range of spectrographically standardised substances is already well established, are now able to supply several graded series of standards for quantitative spectrochemical analysis. These standards have been subjected to very thorough analyses, full details of which are included in the laboratory report which is provided with each set of standards. Publication No. 1763, which can be obtained on request, describes the standards, the ranges of composition that they cover and the type of material which they are suitable for testing.

We have received from Imperial Chemical Industries Ltd. a copy of their latest booklet "Kynal Aluminium Alloys and Fluxes for Brazing." When assessing the suitability of any aluminium alloy for brazing, it is important to consider first the temperature at which incipient fusion occurs, for it is necessary to have a margin of operating temperature of at least 15° C. within which the filler is completely molten. The booklet lists the wrought and cast alloys which may be brazed by the torch, furnace or flux-dip techniques, before giving particulars of the fluxes, filler alloys and brazing-alloy-coated sheet and strip available. The three methods referred to above are described in some detail, together with the cleaning operations required before and after brazing. The design of brazed joints and their properties are also discussed.

LABORATORY METHODS

MECHANICAL • CHEMICAL • PHYSICAL • METALLOGRAPHIC
INSTRUMENTS AND MATERIALS

JANUARY, 1954

Vol. XLIX No. 291

An Improved Method for Routine Electrolytic Polishing of Microspecimens

By R. L. Hancher, A.R.T.C.

A number of variables contribute to the complexity of the electrolytic process for polishing metals. Experiments have shown that the optimum polishing conditions for a particular specimen are dependent on the voltage-resistance characteristics of the cell. In this article the author describes the use of a circuit designed to achieve the optimum conditions easily and quickly.

DURING the last decade, numerous papers have been published on the subject of the electro-polishing of metals, and have included interesting and valuable information. Many of the major problems have been solved, and the industrial application of the technique has proved satisfactory and, in many cases, advantageous. However, it was felt that many of the papers published employed methods not readily applicable to the routine examination of microspecimens, despite the claims made by their authors. It was for this purpose, and to compare some of the more recent electropolishing solutions, that this work was undertaken.

A number of variables contribute to the complexity of the polishing process, although certain industrial and other methods have succeeded in narrowing these down by means of standardisation. Furthermore, it is not easy to reproduce similar conditions for the purpose of comparison. The main variables are as follows:—

- (a) The surface area of the specimen and the depth of its immersion in the electrolyte.
- (b) The current density (for a given current passing through the cell) on the specimen to give optimum polishing conditions is not always the same, but is dependent upon the age and temperature of the electrolyte.
- (c) Specimens having different compositions require different polishing conditions.
- (d) The degree of agitation of the electrolyte is often critical.

Normally the current to be applied to a specimen is calculated by a simple proportioning method. This, however, is difficult with irregularly shaped specimens, and at best rather unreliable.

Film Formation and Polishing

Before describing the apparatus employed to overcome many of these difficulties in a way which provides an almost foolproof method for routine polishing, it would be advisable to consider some of the fundamental aspects of electrolytic polishing, in particular, the curves obtained by plotting voltage applied across the cell

against the cell resistance.* Basically, these curves fall into two distinct types, i.e., those which have a maximum resistance corresponding to a certain voltage, and those in which there is no such maximum. Electrolytes containing perchloric acid as a constituent tend to give curves of the first type (mainly with ferrous specimens), and the method to be described is based on this factor. The electrolytes in the other group are not very satisfactory with ferrous specimens, due to the difficulty of controlling the polishing process, and, consequently, they are not often used in this field.

The functioning of the proposed apparatus depends on the electrolyte having the type of resistance-voltage curve with a maximum. Experiments have shown that optimum polishing conditions for a particular specimen are obtained when the voltage across the cell is such that the voltage-resistance curve has a maximum corresponding to the voltage. This may be explained by considering the polishing mechanism, which involves the formation on the anode surface of a film,¹ which is reddish brown in colour (with steels), very viscous, and is derived from the reaction products evolved during the passage of current through the specimen and the solution. At low voltages, this film does not form, etching and pitting occurring but no polishing. As the voltage is increased, the film begins to form, presumably because the reaction products are formed more quickly than they can diffuse into the bulk of the solution. Further increases in voltage serve to thicken the film, causing an apparent increase in the resistance of the electrolyte (since the film has an appreciable resistance). However, at still higher voltages, oxygen evolution occurs at the anode, and the small bubbles so formed disrupt the film, which in turn becomes unstable and causes an apparent decrease in the solution's resistance. Consequently, the period of maximum stability of the film corresponds to the peak in the voltage-resistance curve.² Obviously then, a knowledge of the conditions of current density, voltage, etc., to give the maximum thickness of anode film, would be very useful in determining the optimum polishing conditions. However, as will be seen later,

* Calculated from voltage and current.

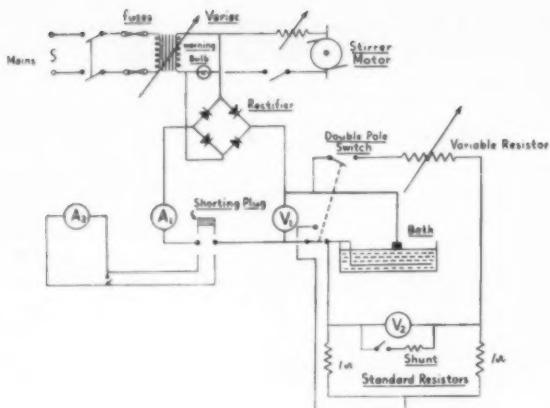


Fig. 1.—Circuit for power supply and balancer.

the conditions are complicated by the variables already mentioned. A graph of resistance against voltage would indicate the conditions under which the cell should be operated to give a film of maximum resistance, and hence the optimum polishing conditions for the particular specimen and surface area exposed to the solution. To plot such a graph for every specimen has obvious disadvantages, especially where routine samples are being polished, and to overcome this difficulty the " Balancer Circuit " was introduced, its prime function being to provide an indication of how the resistance of the bath varies with the voltage applied across it, and, in this way, to facilitate rapid adjustments of current and voltage in the cell circuit to give the optimum polishing conditions for any type of specimen and any surface area.

The Polishing Apparatus

The power circuit is of a conventional design. Supply from the mains is fed via switches and fuses through a Variac transformer. A full-wave rectifier converts the

alternating current to direct current, and this is supplied to the cell. A voltmeter with a range of 1-150 volts, and an ammeter reading 1-10 amps, are included in the cell circuit. Provision is made by means of a plug socket, to insert a more sensitive ammeter when working with certain high resistance electrolytes. When not in use, a shorting plug completes the circuit. A stirrer, with variable speed regulation, and a warning light are also included in the circuit (see Fig. 1).

The assembly is built into the housing illustrated in Fig. 2 to give an apparatus of both pleasing appearance and convenience of design. Further space is available inside the apparatus for extending the range of the instruments to polish larger specimens requiring higher current and voltage. Voltage and current control is effected by manipulation of the Variac transformer setting.

The bath for holding the electrolyte is of $\frac{1}{4}$ in. thick glass, 8 in. long \times 4 in. wide \times 4 in. high, and it sits on the white Vitrolite frontal table of the apparatus (see Fig. 3). A stainless steel cathode, 17.5 cm. long \times 3.5 cm. wide, lies along the bottom of the tank. Alternatively, it can be raised from the bottom by means of tubular glass spacers, thus enabling a greater volume of electrolyte to be carried in the bath when the inter-electrode distance is small. The depth of the electrolyte above the cathode can be read by a centimetre scale inscribed on the vertical side of the bath.

A somewhat unusual anode arrangement is employed. A rigid projection from the control panel carries a stainless steel channel, insulated with bakelite (see Figs. 3 and 4). Two spring clips attached to the channel (which is live and positive) provide a fastening for a nickel rod. This has at one end a rubber hand grip, and at the other a stainless steel clamp for holding the specimens. The object of this design is to enable the operator to remove the rod and specimen for rapid washing. A further advantage is that the rod can be moved up and down in the spring clips, so bringing the specimen into contact with the electrolyte in the bath.

An essential part of the equipment is the cooling arrangement. Glass coils which zig-zag through the electrolyte are used, tap water providing the necessary coolant. Cooling is very important with some solutions, both for ensuring safety and for maintaining the best polishing conditions.

The " Balancer Circuit "

This additional circuit will be recognised as a modified form of Wheatstone Bridge. A large, heavy duty variable resistor is placed in parallel with the cell and the two standard resistors (also heavy duty and about one ohm resistance) are connected as shown in Fig. 1. A 0-5 volt A.C./D.C. voltmeter (with several ranges for different sensitivities) is inserted across the bridge, in the position normally occupied by a galvanometer, in order to accommodate the larger currents passing when the bridge becomes unbalanced at a later stage in the polishing, due to the current heating up the cell and causing a marked lowering in its resistance.



Fig. 2.—The complete polishing equipment.

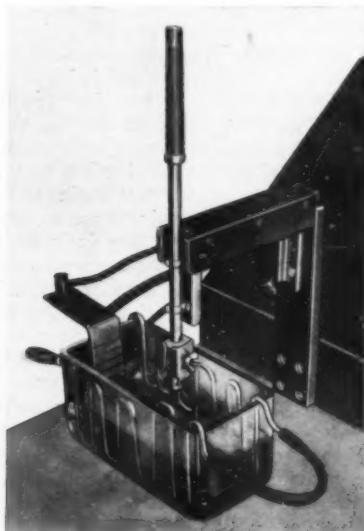


Fig. 3.—The polishing bath.

The operation of this additional circuit is fairly simple. First, the variable resistor is adjusted to approximately that of the cell circuit. This need not be done accurately since the indication of maximum resistance is only a qualitative one, and the variable resistance only affects the sensitivity of the apparatus (the nearer the resistance to that of the cell the greater the sensitivity). The initial setting may be effected by using a resistance equal to that of the cell. Alternatively, a small current may be passed through the circuit and the needle of the bridge voltmeter set to a zero reading by adjusting the variable resistor.

With the cell circuit approximately balanced by the heavy duty resistor, the control handle of the Variac transformer is gradually rotated with a smooth easy movement. It will be observed that the needle of the bridge voltmeter will move across the scale until the maximum resistance in the cell circuit is reached, thereafter the needle falls sharply. However, as the polishing proceeds, the needle will slowly begin to rise again, since the cell resistance, i.e., that of the electrolyte, starts to fall due to the heating effect of the current. Thus the Variac setting must be made fairly carefully, for if it is made too slowly, the maximum as indicated on the voltmeter is masked by the effect of the solution being heated by the current. This is not really a disadvantage, for it is desirable to reach the optimum polishing conditions as soon as possible in order that no pitting, etc., will take place on the surface of the specimen. Furthermore, should the voltmeter fail to record a maximum, the Variac can be turned to give a very low current, and if this is held for a second or two the film dissolves and the setting can be repeated.

Typical Results

A large number of different ferrous specimens were polished using this attachment in conjunction with the perchloric acid solutions, and excellent results were obtained. A number of non-ferrous specimens were also polished, using the same procedure, and here the equipment appears to have even greater scope, since most of the electrolytes used for non-ferrous work give

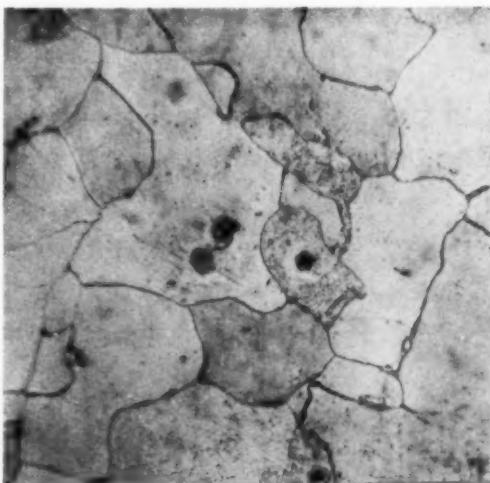


Fig. 5.—Armco iron, polished from 000 emery paper in 20 seconds, using Modification A of the de Sy and Haemer Solution. Etched in bath at low current density. $\times 300$

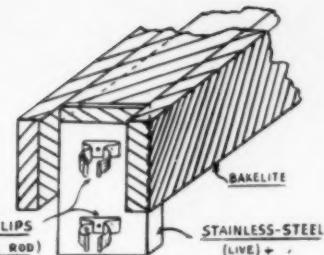


Fig. 4.—Details of arm for holding anode.

graphs of resistance-voltage with definite maxima. (See Figs. 5-10).

Several solutions were used for these experiments, all giving good results. The electrolytes were as follows:—

I—Jacquet's New Solution.

Perchloric Acid	50 ml.
Glacial Acetic Acid	1000 ml.

II—The de Sy and Haemer Solution.

Perchloric Acid (70%)	36 ml.
Ether	16 ml.
Water	97 ml.
Ethyl Alcohol	500 ml.

III—The de Sy and Haemer Solution—Modification A.

by weight	
Perchloric Acid (sp. gr. 1.2)	2 parts
Glycerine	1 part
Ethyl Alcohol (96%)	7 parts

IV—The de Sy and Haemer Solution—Modification B.

by weight	
Perchloric Acid	2 parts
Butyle Cellosolve (ethylene glycol-monobutyl ether)	1 part
Ethyl Alcohol	7 parts

These solutions are all eminently suitable for use with the "Balancer Circuit" although Solution IV is probably

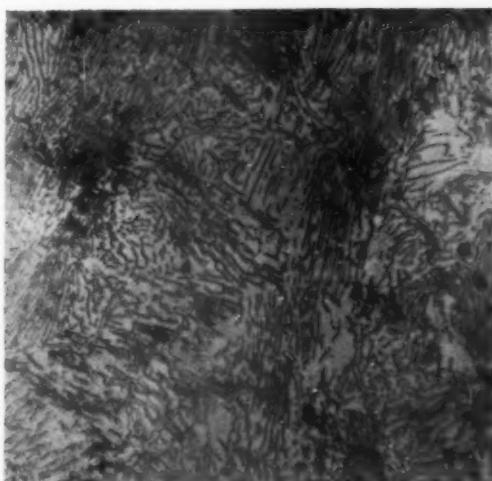


Fig. 6.—1.2% carbon steel, polished from 1 M emery paper in 45 seconds, using Modification A of the de Sy and Haemer Solution. Etched in 2% nital. $\times 700$



Fig. 7.—Badly overheated and burnt 0.25% carbon steel, polished in Modification B of the de Sy and Haemer Solution. $\times 75$

the best as regards speed and quality of the polish. Non-metallic inclusions are also rather less attacked by this electrolyte. A typical polishing time for this solution from a 000 finish is 30 seconds or less. Solution I also gives excellent results, but for a similar initial preparation of the surface, the time required for polishing is of the order of 2 minutes.

The original Jacquet Solution can also be used with the "Balancer Circuit," but, due to its explosive nature and the long time needed for polishing, the other solutions are preferable, giving as good or better polishes with a maximum of speed and safety.

"Star" Effect

This phenomenon was observed principally with the de Sy and Haemer Solution (and modifications) and to some extent with Jacquet's New Solution. The effect is



Fig. 8.—White cast iron, polished in Jacquet's New Solution. $\times 60$

not readily noticed when the specimen has been etched, but can be more easily seen on the bright unetched surface. Viewed through the microscope, the effect takes the form of clusters of bright lines radiating from a common centre to give star-like shapes. Any number of these lines may be found in a "star," and the "stars" occur to a greater or lesser extent all over the surface. The lines do not appear to be very deep in the surface, and are generally removed on etching. The locations of the "stars" are seemingly random, and do not correspond to the outlines of grain boundaries, etc.

An increase in the degree of stirring often removes the "stars," but this is not always positive and they often appear quite unexpectedly. The electrolyte with which this effect is most noticeable is Modification A of the de Sy and Haemer Solution. With the others it occurs to a lesser extent; hardly at all with Modification B.

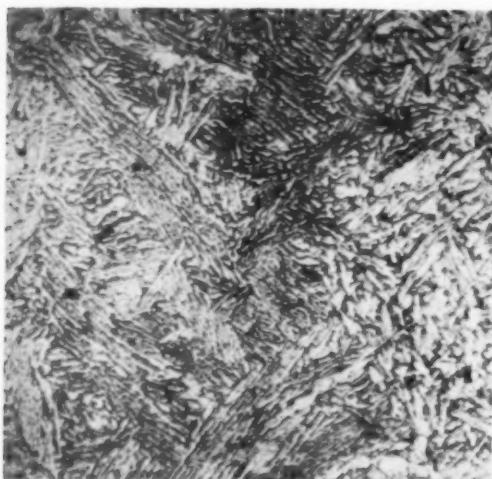


Fig. 9.—60:40 brass, extruded below the recrystallisation temperature, polished in Modification A of the de Sy and Haemer Solution. $\times 100$

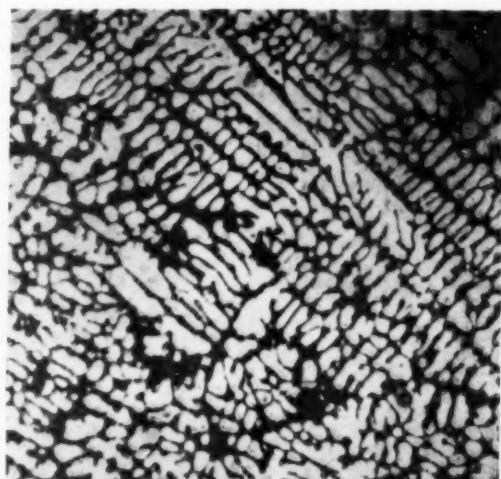


Fig. 10.—80:20 aluminium-copper, as cast, polished in Modification A of the de Sy and Haemer Solution. $\times 60$

Effect of Steel Composition

A number of ferrous specimens of different carbon contents were used. These were all normalised specimens and all had the same surface area and the same depth of immersion in the solution. Figures were obtained for the voltage applied across the cell and the corresponding current density, and from these, graphs of voltage against resistance for each of the specimens were plotted (Fig. 11).

The main feature illustrated by this graph is that there is considerable variation between the voltages at which the most stable film forms on the different specimens. When these are plotted against the carbon content, a curve is obtained, which indicates that the maximum voltage corresponding to maximum resistance, occurs with the eutectoid steel. However, the carbon content may be incidental, i.e., controlling the hardness of the specimen and perhaps its ease of solution in the electrolyte, thereby influencing the voltage at which the most stable film forms.

The most important feature that these curves reveal, is that the polishing conditions vary for different specimens and that the optimum conditions cannot easily be estimated.

Effect of Age of the Electrolyte

It is difficult to give an estimation of the life of an electrolyte in terms of the number of specimens which can be polished, for this is complicated by the fact that, whilst graphs, etc., are being made, the specimen is in contact with the solution for much longer periods than those required for normal polishing, thus making the assessment harder. However, in the case of some of the electrolytes used, the resistance of the new solution may have been of the order of 250 ohms, whereas, after a number of specimens have been polished, the resistance under the same conditions (i.e., same specimen, inter-electrode distance and surface area) drops to below 60 ohms. This is due to an increase in the conductivity caused by increasing the metal content of the electrolyte. The solutions also turn a dark brown due to the ferric

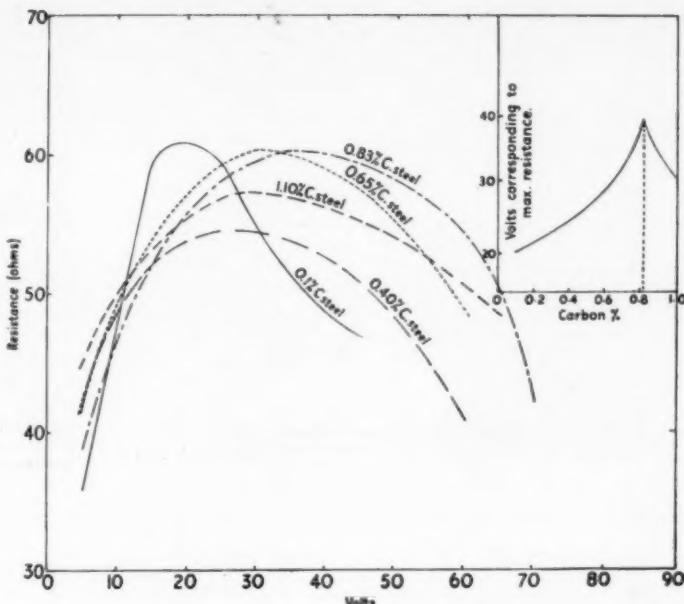


Fig. 11.—Voltage-resistance curves for steel specimens of varying carbon content. Inset: Voltage corresponding to maximum resistance plotted against carbon content. All these were normalised steels and had the same surface area.

ions. In addition, the voltage corresponding to the maximum resistance falls as the metal content increases. This again is another advantage of the "Balancer Circuit," which is independent of voltage changes due to the age of the electrolyte.

It will be appreciated, therefore, that a bridge circuit of this type can remove many of the uncertainties associated with the routine polishing of specimens by electrolytic methods. If properly operated, any specimen can be polished under optimum conditions, so ensuring the best possible finish.

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Certification Scheme for Samples for Metallurgical Analysis

IN connection with B.S. 1548; 1953—Certified Samples for Metallurgical Analysis—the British Standards Institution has just published an outline of the general procedure to be adopted. This is as follows:—

Application.

(1) A manufacturer of reference samples for metallurgical analysis applies to B.S.I. for permission to use the certification mark. He describes in detail: (a) the material for which a licence is asked; (b) the method by which it is proposed to prepare the bulk sample to secure homogeneity; (c) the method of taking test samples from the bulk sample; (d) the total quantity of material available for test samples (to secure supply over reasonable period); and (e) the method of packing and storage of samples.

(2) The applicant will at the same time agree to accept inspection at any reasonable time by a representative appointed by the B.S.I. to comply with the scheme of supervision and control, and to pay, up to an agreed maximum annual figure, the costs of inspection, check analyses by B.S.I., and administration.

Mark Sub-Committee.

(3) The Mark Sub-Committee considers the application and approved it in principle, either as it stands or subject to such changes in procedure as may appear necessary.

(4) The Sub-Committee appoints (a) independent experts to advise upon the method of preparation of the bulk and test samples; (b) an inspector to visit the applicant's works and report; and (c) six analysts to

make the test analyses and furnish analysis certificates.

Any of the above may be appointed either among the members of the Sub-Committee or otherwise.*

Inspections.

(5) The inspector visits the applicant's works and reports fully on his equipment and the conditions of preparation, analysis, packing and storage of samples.

(6) He satisfies himself that the agreed method of preparing samples to secure homogeneity is followed, and takes or supervises the taking of samples for despatch to the six analysts. He selects or supervises the selection of a reference sample which is sent to B.S.I. for purposes of later comparison with samples currently on sale. He may also supervise the operation of bottling and storing samples for sale.

Analysis.

(7) The six analysts having agreed to act forward particulars of the method of analysis they propose to follow (see Clause 3, B.S. 1548). They make the analyses and forward their certificates of analysis to B.S.I.

* Any purchaser may obtain from the B.S.I. particulars of the experts, inspectors and analysts appointed in respect of any certified sample.

Approval and licensing.

(8) The Mark Sub-Committee inspects and approves the certificates and recommends the issue of a licence to use the mark.

(9) The B.S.I. issues its formal licence to use the certification mark, accompanied by a certificate giving the results and methods of analysis.

(10) The licensee attaches to each sample sold the B.S.I. certification mark, his licence number, the batch number and a copy of the certificate of analysis issued with the licence.

(11) When the licensee proposes to take a subsequent batch of samples from the same bulk he applies for endorsement of his licence. The procedure described above is followed as may be directed by the Mark Sub-Committee, but in such cases a minimum of two analyses may be accepted. Samples from the second or later batches bear the licence number followed by /2, /3, etc.

(12) The B.S.I. buys at least one certified sample in the open market every two years and has a check analysis made on this and on its retained reference sample. The result is reported to the Mark Sub-Committee.

Advisory Service to Electroplaters

SINCE its early days, The British Non-Ferrous Metals Research Association has accepted the responsibility of helping industry in the application of scientific knowledge to its production processes. This has not been confined to the exploitation of the Association's own researches, although that, of course, has been one of the major objectives. The work is carried out by the Liaison and Technical Service Department established originally many years ago under the name of the Development Department, and now responsible for over 20% of the total expenditure of the Association.

From the substantial sums of money existing as the sterling counterpart funds created to balance the dollar funds made available to this country by the U.S.A. under the Mutual Security Act of 1952, H.M. Treasury is making grants to finance projects intended to aid the economy of this country by promoting industrial productivity. An expansion of the existing machinery of the Association to carry out work of this type is a fairly obvious step, and the Association has, in fact, received an offer of up to £15,000 to be spent with this object within the next three years. A substantial part of this sum is to be expended in the provision of an advisory service to electroplaters.

Productivity in the Plating Industry.

The Research Association, in consultation with members of the Metal Finishing Association and with other companies in its membership having plating interests, is undertaking a scheme whereby a small team of investigators visit plating shops by invitation to study methods of working, plant layout and other factors concerned principally with the technical aspects of plating practice. Observations made in a considerable number of plating shops of different types are being studied, and the conclusions reached affecting matters of good practice in achieving high productivity will be made available in due course to firms with plating interests.

At the same time, the individual firms co-operating in the initial survey are advised of those features of their practice where improvements could be effected practically and economically. The interests of the firms providing the information on their operations are fully safeguarded, and the sources of information will be known only to the Research Association's staff. If the initial survey indicates that it is likely to be useful, supplementary questionnaire surveys of somewhat wider scope will be undertaken.

In parallel with this field work, the Association has set up a pilot scale plating line which operates on a continuous production basis. This plant embodies all the good features of current plating practice, and will serve to demonstrate methods of operation necessary for the consistent production of high quality plated coatings, so increasing productivity by reducing waste.

Royal Society Officers

THE following have been elected Officers and Council of The Royal Society for the ensuing year—

President : Dr. E. D. Adrian, O.M.

Treasurer and Vice-President : Sir Thomas Merton

Secretaries and Vice-Presidents : Sir Edward Salisbury, C.B.E. and Sir David Brunt.

Foreign Secretary : Sir Cyril Hinshelwood

Members of Council : Dr. F. P. Bowden ; Professor F. Dickens ; Professor H. J. Emelius ; Mr. A. E. Ingham ; Dr. G. M. Lees ; Professor E. G. T. Liddell ; Professor R. A. McCance, C.B.E., Professor P. B. Medawar ; *Vice-Presidents*, Sir Edward Mellanby, G.B.E., K.C.B. ; Professor P. B. Moon ; Professor W. H. Pearsall ; Professor R. O. Redman ; Professor Alexander Robertson ; Dr. C. Sykes ; Professor G. Templer Professor C. M. Yonge

THE B.T.H. Middlesbrough Office is now at Prudential Chambers, 27, Albert Road, Middlesbrough. The telephone number remains unchanged at 2476.

